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Crystals of primary cementite, Fe_3C. Half natural size. §195.

Crystals of primary austenite. Half natural size. Vickers, Ltd. §§187, 343.

(*Frontispiece*)

THE METALLOGRAPHY OF STEEL AND CAST IRON

BY

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"A HAND
CONDUCTS ME THROUGH THE CLOUD ROUND LAW TO WHERE I STAND
FIRM ON ITS BASE,—KNOW CAUSE, WHO, BEFORE, KNEW EFFECT."
Browning, Ivan Ivanovitch

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To

HENRI LE CHATELIER

MEMBER OF THE INSTITUTE OF FRANCE,

PHILOSOPHER, TEACHER, LEADER,

AS A TOKEN OF AFFECTION AND ESTEEM

THIS WORK IS DEDICATED

PREFACE

This volume, to which I hope to add a series of others, consists of two distinct parts, an introduction to the new science of microscopic metallography, as applied to steel and to cast iron, and an extended study of the very new branch of that science, the mechanism of plastic deformation. It is through the application of the principles of these subjects that the great advances in the metallurgy of iron and steel are to be expected.

The mechanism of deformation has been before us so little that a word as to its importance may not be amiss. The usefulness of steel really results from its resistance to deformation, and its power to endure limited plastic deformation. Hence a knowledge of the mechanism of this deformation and of the way in which steel in part resists deformation and in part accommodates itself to it, may in time disclose to us the essence of its power of resistance and accommodation. To understand this essence is to be the better prepared to approach the problem of fitting the metal for its service to our race, not empirically alone but also scientifically.

As befits such an attempt as this, though I try to explain and illustrate clearly the visible phenomena, my chief aim is to stimulate others to think profoundly, in order that some among them may in due time push discovery farther and farther beyond its present very early stage. Those of us who are working and thinking on this subject today are only crying in the wilderness, in the hope of inciting our successors.

With this general aim I have acted on Tyndall's sage words:

"Right or wrong, a thoughtfully uttered theory has a dynamic power which operates against intellectual stagnation; and even by provoking opposition is eventually of service to the cause of truth."

The true task of the teacher is to excite thought. Hence I do not hesitate to offer such hypotheses as I can devise, not with the belief—hope should not enter into consideration—that they will endure, but with the aim of stirring others to seek the truth by destroying them. This, I take it, is the true function of most hypotheses, and this purpose should be in the heart of every philosophic student. Laws are useful, and where opportunity offers I try to deduce them. But because the causes beneath those laws are of a higher order of importance, it is to these that I have chiefly addressed myself.

Greatly as steel and cast iron differ from each other in manufacture, properties, and application, they form for the metallographist two conterminous divisions of a single series, so closely united that, for the purposes of a general survey, they may well be studied together. Partly on this account, and partly because metallography seems to me to hold out even greater promise to the iron founder than to the steel maker, I have devoted a con-

siderable fraction of this work to the metallography of cast iron. I find strong reason to hope that the intelligent application of metallography may succeed in improving the properties of cast iron to such a degree as to broaden its application greatly. Indeed, a relatively small degree of improvement might well lead to a disproportionate broadening. That so little has been done intelligently to improve the properties of this product by applying metallography to it can be explained in part by the extreme complexity of this branch of the subject, and in part by the usually slighter technical training of the iron founder than of the steel maker. But that so little has been done, far from disproving, tends rather to suggest that much remains to be done, and to add to the attraction of this field, as yet only roughly surveyed.

It is with profound gratitude that I acknowledge my indebtedness to many friends. First to my assistant, Mr. Arthur G. Levy, for his untiring, zealous, and skillful coöperation, especially in the many experiments made to test my hypotheses, and in passing the work through the press. Then to the stimulating encouragement, advice, and unfathomable fund of knowledge of Dr. John Edward Stead, F. R. S. My colleagues, Professors William Campbell and Alfred J. Moses, have aided me with their wide knowledge of metallography and of crystallography, and Professor James S. Macgregor has kindly made the tensile tests for which the work has called. Messrs. Booth, Garrett, and Blair have generously made most of my chemical analyses, Mr. Wheaton B. Kunhardt, Treasurer of the Carpenter Steel Company has given much invaluable material for experimental purposes, and many others have helped me in this way, notably Dr. John A. Mathews, Managing Director of the Halcomb Steel Company, Dr. George W. Sargent, Metallurgist of the Crucible Steel Company of America, the late Fred H. Daniels and Mr. Clinton S. Marshall of the American Steel and Wire Company, Mr. James A. Aupperle of the American Rolling Mill Company, Sir Robert Hadfield, F. R. S., and others whose names are given in Table 39, p. 616.

H. M. H.

Green Peace,
December, 1915.

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THE METALLOGRAPHY OF STEEL AND CAST IRON

CHAPTER 1

INTRODUCTION

1. Introductory Survey.—Iron, including those species of it which are called steel, is the most abundant and hence the cheapest of the metals; the strongest and the most magnetic of known substances; the most abundant of man's earthly possessions if we admit that the earth is a huge iron meteor with but a relatively thin covering of rocks; and perhaps also the most indispensable of all, save the water we drink and the air we breathe. These two substances alone seem absolutely indispensable to our existence. For any one kind of meat we could substitute another; indeed the vegetarians show us that meat is not a necessity, and may be a curse. If we lacked our common silicate glass, we could probably perfect some other transparent solid to take its place. For wool or indeed any one of our textile substances we could substitute the others, or fur. Cotton, hemp, and endless fibrous vegetable substances await our convenience. But even if we could compel the earth to yield us any convenient substitute for the forty or fifty million tons of iron that we use each year for vessels, rails, wire, buildings, bridges, machinery, and agricultural and domestic implements, we do not now see how we could replace iron for our cutting tools, or, except at very greatly increased cost, for our magnets, which are the basis of all commercial electricity, and of the telegraph and the telephone.

The indispensableness of iron is due to a cause radically different from that of the indispensableness of air and water. These two are indispensable because their ubiquitousness has led to the evolution of an order of things absolutely dependent on them. Man and his earth-born companions have ever been the slaves of their environment. It is this environment which has determined the path of evolution.

But the abundance of iron has not determined the evolution of man, because it was not until man had already been evolved as man, perhaps not till he had risen from savagery into barbarism, that he began to use iron. Till then his evolution had gone on unaffected by the abundance of iron. Only then did he begin to be its master; it was long after, some may say that it is only in our day, that he has become its slave.

Iron owes its usefulness first to certain single properties in which it excels all other substances; its strength, its magnetism, and its hardening power, *i.e.*, the power of shifting readily to an intensely hard state from a

soft and malleable one, as in the hardening of steel tools by quenching in water; second, to certain combinations of properties in which it excels, such as its strength and hardness when cold with its ready welding and shaping when hot; and third, to the astonishing diversity of the properties of its different varieties, a diversity due chiefly to the great influence of the carbon which iron always contains, though in very varying proportions, and to the allotropy of the metal itself, that is, to its having radically different states between which there is a difference like in kind though much less in degree than the difference between diamond, lampblack and graphite, the familiar allotropic states of one element, carbon. Thus it comes that iron is a veritable Proteus. Here it is extremely magnetic, there almost nonmagnetic; here it is as brittle as glass, there almost as ductile and soft as copper; here it is most resilient, there it lies almost as springless and dead as lead; here it is readily fusible, there very infusible; here it welds readily, there it can hardly be welded at all; here it is an excellent conductor of heat and electricity, there it opposes their passage stoutly; in our files and razors it is extremely hard, in our horse-shoe nails it is so soft that the smith rejects them unless he can bend them on his forehead; here it is so strong as to be by acclamation the metal of Mars, there contemptibly weak.

To those who ask whether design underlies evolution, the extraordinary abundance of this most indispensable substance, awaiting the day when other agencies shall have evolved a being capable of using its hidden properties, of drawing the sword from the tree, is certainly suggestive. Is it an accident that iron-needing man was planted on an iron-cored globe, and evolved without the aid of iron to a state in which he could use to the full this metal which is his most abundant and now his indispensable tool? Be our surmise what it may, can it rank with the minnow's conception of the thoughts of the philosopher standing on the bank? Can we finite creatures gauge the Infinite as well as a duller finite being can gauge a keener one? Vanity aside, in which of these cases is the disparity between the judge and the judged the greater?

2. Historical Sketch.[1]—Lumps of iron oxide in the form of iron ore are so widely scattered over the face of the earth, and their iron so easily reaches the metallic state by yielding up its oxygen to glowing carbon, that metallic iron must have been made unwittingly in the camp fire, very early in man's ascent through the lower stages of savagery. The resultant lumps of metallic iron must early have caught his notice by their heaviness, recalling the heaviness of the ore out of which they were formed; by their white metallic look where they were bruised; and by the dents which shone in these white bruises, testifying that this substance differed most preciously from all other hard or heavy or strong things then known, in that it could be dented

[1] On the history of iron and steel compare Beck, "Geschichte des Eisens," 1884; Ledebur, "Handbuch der Eisenhüttenkunde," vol. 1, p. 8, Leipzig, 1906; Swank, "Iron in All Ages," Philadelphia, 1892 (Second Edition), and the article on "Iron and Steel" by the late C. Alder Wright in the ninth edition of the "Encyclopædia Britannica," vol. 13, pp. 278 to 359.

without breaking, in short that it was malleable. Slight as was the suggestion of malleableness which these dents made, it was a suggestion of something in which the iron, so unwittingly made, differed not in degree but in kind from all known hard and strong things. And from all known things it further differed in its great heaviness and in the metallic whiteness of its bruises—from all known things unless copper, gold, or silver had already become known, in which case the heaviness and malleableness of the iron would tend to class it with these metals, and help to suggest that this malleableness might be made useful.

By and by some savage, long familiar with these signs of malleableness, and helped, be it by occasion, be it by constructiveness, would become the father of the iron industry by putting some favorably shaped lump to use, to pry, to rip, to throw, or to pierce. In a sense the first iron metallurgist proper as distinguished from the first iron user was he who first purposely made iron thus, in another sense he was the very first who made it, centuries before, without even noticing it.

3. Independent Discoveries of Iron.—Like causes leading to like effects, and the means of diffusing information being extremely scanty, it is antecedently probable that the use of iron originated in many places independently. The savage in South Africa, for instance, whose circumstances happened to direct his steps into a sequence like that just sketched, would be very likely to get some use out of these metallic lumps before the recipe for their preparation reached him from India, for instance, by the slow process of diffusion across two continents. In this view the primitive iron industry grew in part by diffusion, in part by independent initiation.

It is true that the general similarity between the primitive iron extraction of places which, though far apart, yet might possibly communicate with each other because they are all in the eastern hemisphere, coupled with the apparent ignorance of iron on the part of so advanced a people as the Aztecs in the western hemisphere, afford some comfort to those who hold that the use of iron originated in one place only. But first this similarity might well result from the necessary conditions, the much greater ease of this one line of development than of any other. And next the argument from the Aztecs' ignorance comes near to saying "Because there is one people that failed to discover the use of iron, all must have failed save the first."

Independent discovery of the use of iron might be especially easy where there was meteoric iron.

If it should be shown both that great numbers of tribes in the eastern hemisphere, very widely scattered and on a lower plane of barbarism, had closely similar archaic iron methods, and also that, in spite of there being many peoples on a higher plane and with iron ores accessible in the western hemisphere, not one of these had used iron at all, the argument for a single origin would certainly be attractive.

Once any use of this strong heavy substance was known, use would follow use, supply would follow demand, and getting iron from its ores and shaping it, iron metallurgy in short, would become an art and trade.

4. Early Discovery of Steel.—Much of the early craftsman's product would inevitably absorb so much carbon from the coals by which it was reduced from the ore that it would have the "hardening power," the power of being made hard or soft by rapid or slow cooling from a red heat. In short it would be steel in the historic sense of that word. By and by this iron smelter would learn that he could shape the metal hot more easily than cold, perhaps pounding it with a hand stone as it lay on a large flat rock as an anvil. But, once he had brought it to shape he would wish to cool it, to save both time and the burning of his own and his visitor's fingers, and in order to cool it, he would push it into water. When he had thus cooled a piece of this steel from a red heat, he would find that he had replaced one most useful property with a strikingly different if equally useful one, malleableness with intense hardness and great strength. Perhaps he then began that discussion about the hardening of steel that has retained to our own day the hotness of the art itself, and a ferocity worthy of Mars[1] the patron of the metal.

Among the very first things in the progress of the craft would be putting these metal-yielding stones intentionally into the fire, adapting the fire to them instead of fitting them to the fire. And as the way in which, in the struggle of the elements, wind is the ally of fire even in its fight against water must long have been a matter of common knowledge, the craftsman might soon come to putting his fire in a hole open to the prevalent wind, making a forge like in kind to the "Hornos a gran tiro" of our own time, and differing from the Catalan forges of today only in size, in lacking forced draft, and in details of construction.

5. Relative Age of Iron and Bronze.—Copper is reduced from its ores so much more easily than iron and then is so much more striking in color and in malleableness—indeed the brilliant green, blue, and red of the ores of copper are so much more striking than the more somber hues of iron ore—and native copper is so much more noticeable than meteoric iron, that, where either copper or its ores existed along with iron or its ores, Venus was likely to outwit Mars, as she has ever since, and bring her metal the earlier into use. Thus in the Homeric poems bronze is far more prominent than iron; but this tells us only of a single place and time.

If the use of each metal originated in various places independently, then iron was probably used before copper in some places which had rich iron ores but lacked those of copper, and copper before iron in other lands where these conditions were reversed.

Of course at some one hour and spot man for the first time used either of

[1] Iron was called "Mars" by the alchemists, and its peroxide formed by calcining the hydrate or ferrous carbonate in the air was called "crocus Martis" till relatively lately. Even as late as 1875 "Watts's Dictionary of Chemistry" gave this name as then current (vol. 3, p. 394).

these metals; but which of them it was we can never know, and any supposed order of precedence must have reference only to a certain specific place. It is striking that even today the traveler finds primitive folk who extract iron from its ore and use it, but know nothing of copper.

If iron thus preceded copper in some places, in still more must it have preceded brass and bronze, those alloys of copper with zinc and tin respectively, for neither of them would be made so often and so readily as copper or iron, and hence they could not press as copper and iron pressed for recognition and adoption.

After iron had come into use, the great ease with which the copper alloys can be cast into useful shapes may have led to their replacing it in large part, and perhaps wholly in certain places, if the primitive are such slaves of fashion as their descendants. But because the far greater hardness of steel than of these alloys is of such great value even to the primitive, it seems in the highest degree improbable that this displacement of iron can ever have been universal. We are not to be misled by talk of the lost art of hardening bronze, nor by the absence of iron from among prehistoric remains rich in bronze objects. That the ancients should have stumbled on an art of hardening bronze which the searchlight of today's science fails to detect is improbable enough antecedently; but quite apart from this, if such an art existed, then many of the abundant bronze relics of antiquity ought to be hard, but they are not. Again, the absence of iron relics should be accepted as evidence only with great caution, because the exasperating proneness of iron to rust, whether this is the penalty for the blood it has shed or not, might easily lead it to disappear even though bronze objects once laid down beside it were but little changed.[1] Indeed, in view of the rapidity with which iron rusts away and vanishes, the presence of such iron objects as antiquaries actually have found may be likened to the single foot print of the savage which the waves left to warn Robinson Crusoe.

6. Great Antiquity of Iron.—Now that we have seen how naturally, indeed inevitably, primitive man would first come to recognize and then to use iron, and in using it would stumble on the hardening of steel, we hardly wonder at the hoary antiquity of our craft; we hardly wonder that Genesis credits Tubal Cain, the sixth in descent from Adam, with using iron; that the Assyrians had knives and saws which could hardly have been useful unless they were of hardened steel, *i.e.*, of iron which had become carburized in its reduction from the ore, and had then been hardened by quenching in water

[1] The usual life of buried but unprotected steel and wrought iron piping is about 40 years (F. N. Speller of The National Tube Co., private communication, Sept. 30, 1914). In sinking a caisson in the East River, New York City, a wrought-iron anchor was found which must have been buried for more than 87 years. Its shank had been reduced from an initial thickness of about 2½ in. to between 2 in. and ¾ in. (Walter E. Spear, Dept. Engineer, City Aqueduct Dept., New York City, private communication, October 8, 1914). A circular from the A. M. Byers Co., of Pittsburgh, in 1913, describes one of their 6-in. black pipes which, after remaining buried for 28 years, had lost not more than 3 per cent. by weight. The thread was still in good condition. I have myself examined this pipe. (See *Iron Age*, vol. 90, July 4, 1912, p. 35.)

while red-hot; that the ancient pyramid of Kephron, built probably as early as 3500 B.C., had imbeded in it an iron tool; that by about 1500 B.C. the subjects of Thotmes III were such adepts in iron working that they used bellows for forcing the smith's fire; that in the Homeric age, not later than 900 B.C., the art of hardening steel was so widely known that the poet drew from it a simile, likening the hissing of the stake which Ulysses drove into the eye of Polyphemus to that of the red-hot steel when the smith quenches it,[1] and referring to the strengthening effect of this treatment; that iron tools are referred to early in the Bible;[2] nor that Pliny (A.D. 23 to 79) should have known something of the relative value of the different liquids for quenching steel, and should have preferred oil for hardening small tools.

7. Three Great Periods.—The story of the progress of iron metallurgy may be the more easily told if we split it into three periods: the primitive in which only wrought iron and steel extracted directly from the ore were used; the cast-iron period in which these primitive forms were supplemented by cast iron, used either as such or for conversion into wrought iron or steel; and the Bessemer period, which gave us a temperature high enough to free even wrought iron from its enclosed slag by melting it. The first period, as we have seen, began in some early stage of savagedom; the second began in the fourteenth century, and the third when Bessemer invented his process, about 1855.

8. The Primitive Period.—We can fancy how with painful slowness, compared with which the crawling of the glaciers is an impetuous charge, the savage by Darwinian steps bettered his mastery over his fire, then as so often now both his heating and his deoxidizing agent. When he missed the aid of the wind he would first blow with aching cheeks, then urge his fire with his fan, and after centuries would make the earliest bellows, valveless of course, by sewing the skin of a goat or pig into a bottle, with one leg forming its neck. Such bottles are still used in the East for watering the streets. "I am become as a bottle in the smoke" says the Psalmist, aptly enough as regards such a leathern one. The metallurgist may thus have made the very earliest of all engines, inflating the skin by pulling its sides apart, and blowing the blast out by pressing or walking on it.

[1] "And as when a smith dips an axe or adze in chill
water with a great hissing, when he would temper it—
for hereby anon comes the strength of iron—even so
did his eye hiss round the stake of olive." *The Odyssey of Homer*, IX, 371–402, by S. H. Butcher, M. A. and A. Lang, M. A., New York, 1906, p. 148.

[2] II Kings, Chap. 6, verses: "5. But as one was felling a beam, the axe head fell into the water: and he cried, and said, Alas, Master! for it was borrowed.

"6. And the man of God (Elisha) said, Where fell it? And he showed him the place. And he cut down a stick, and cast *it* in thither; and the iron did swim."

(American Bible Society, 1900 edition.)

Correcting the dates given by Dean Milman, this seems to have taken place between 849 and 836 B.C. According to the "Encyclopædia Britannica," Eleventh Edition (Article, "Bible," vol. 3, p. 852) this narrative was probably written before the Babylonian captivity, B.C. 586.

To make this inflation easier he would later pull the sides apart with strings. To avoid drawing the ashes back into the bellows on thus inflating them, he would complicate and better his engine by using one of its legs as an inlet and another as an exit port, pinching together the sides first of one and then of the other leg-hole.

To these crude bellows some forgotten Watt added valves in the fourth century A.D. Counting back from this time to 1500 B.C. when valveless bellows were already in use, we may get some glimmering of the sluggishness of metallurgy's early advance, if we remember first that progress in this art like all progress must have undergone nearly continuous acceleration, and second, that this 2000-year step from using bellows to giving them valves was taken in a relatively advanced stage of development. If this was the rate when progress had already undergone great acceleration, what must it have been in Tubal Cain's day? If we compare with this the rate of progress which in the lifetime of many now living has given man the locomotive, the telegraph, the telephone, the electric motor, and with bewildering swiftness the Röntgen rays, wireless telegraphy, the journey to the poles, aviation, and the end of the last important absolute monarchy, can we either give fit thanks for the privilege of living in this age or conceive the "skiey speed" of far future progress which this acceleration, continuing, is sure to cause?

It is in keeping with the slow progress of those days that till the fourteenth century the crude forges in which the smith, the "Waldschmied," made iron from his local ores and shaped it, were most wasteful of time, labor, fuel, and iron itself, and in general but little better than those of a thousand years before. Indeed many farmers made their farm tools in their own little smithies.

The most wonderful relics of the primitive period are the wrought-iron pillars of Delhi and Dhar, in India, made about 300 A.D. The Delhi pillar, which is more than 16 in. in diameter and 22 ft. long, seems to have been made by welding together blooms each weighing about 80 lb. It still stands firmly upright. Indeed it is said to have resisted the impact of cannon shot. Still greater is the Dhar pillar. The combined length of its three existing fragments is 42 ft.[1]

9. The Cast-iron Period. Indirect Process of Making Wrought Iron.— Much of the iron thus extracted from the ore would have not only enough carbon to be true hardenable steel, but even enough to be true cast iron, and hence so brittle as to be of little value. Whether through using these lumps of cast iron for building the sides of his forge, or through believing that they were underdone instead of overdone and hence throwing them back into his fire, the smith would some day have it forced on him that, when he happened to remelt these brittle lumps in such a way that they were exposed to the blast, he made them malleable, as we now know by burning out their excess of carbon. Thus must have begun the gradual replace-

[1] Hadfield, *Journ. Iron and Steel Inst.*, 1912, vol. 85, No. I, p. 153.

ment of the direct by the indirect way of making iron from the ore, a replacement which has only barely completed itself in our own time. The smith hastened it unwittingly by making his forge larger and higher so as to work faster and cheaper, and by using water power for driving his blast. Indeed it was by this use of water power that he was able to drive the blast through the longer column of fuel and ore in his heightened forge. But this very heightening of the forge, and so lengthening and strengthening the deoxidizing and carburizing action, caused more and more of his product to become cast iron, which he had to remelt and decarburize in order to make it malleable.

In such a heightened forge the smelter would not only make cast iron but melt it, and because of this melting the wrought iron or steel later made from this product would be very much better than any known before. The reason for this is that the ores of iron as they occur in the earth's crust are mechanically mixed with barren minerals, and in bringing the iron of the ore to the metallic state without fusion, as had been the case in all previous work, relics of these barren minerals remain in the form of "cinder" or "slag" mechanically mixed with the iron and, by breaking up its continuity, weaken it greatly. But as soon as the cast iron is thus melted the light slag immediately removes itself by rising to the surface of the heavy molten metal, as oil rises from water.

When this cast iron so purified was later turned by a decarburizing melting into wrought iron, this product would have, in its freedom from this initial barren slag, a merit so great as to outweigh the other advantages of the high forge, the saving of time, labor, and fuel. Here we have an example of the familiar type of the replacement of the simple but wasteful by the complex and economical, and of a change made for one purpose leading to an unexpected and far greater good, one so great as to determine the path of future evolution.

10. Even this Wrought Iron was Slag-bearing.—We should recognize that, though the wrought iron and steel thus made were free from the relics of the barren minerals of the ore, yet they were contaminated by another impurity, slag or cinder in the form of iron oxide or iron silicate. The reason for this is that, in oxidizing the carbon of the cast iron in order to convert it into steel or wrought iron, much iron also oxidizes. As long as the metallic iron remained molten, this oxide escaped by rising to the surface; unfortunately before enough carbon was removed to complete the conversion into steel or wrought iron, the metal became so infusible, because so poor in carbon, that it solidified in the form of pasty particles, and these particles inevitably retained mechanically much of the iron oxide which formed during the further oxidation and removal of carbon. The indirect method of making wrought iron and steel, then, though it purified them from the relics of the barren minerals of the ore, yet left them contaminated with iron oxide formed in the purification process itself. It was a palliation, not a complete removal of this weakening foreign matter.

11. Iron Founding.—As the art of brass founding was invented relatively early, it is not surprising that the molten cast iron so made was early cast into shapes in which it was useful in spite of its brittleness, into ornamental castings in Sussex in the fourteenth century, and into cannon weighing as much as 3 tons in the sixteenth.

The indirect process naturally developed itself along the line of obvious economy, by making the furnace higher and higher, as the simultaneous advances in mechanical engineering put stronger and stronger blowing engines at the disposal of the iron master, for forcing his blast up through the higher column, in spite of the greater frictional resistance which its height caused. This increase in the size of these furnaces, still called in French and German "High furnaces," has been going on ever since, though of late years the increase has been in diameter and not in height.

12. In the fifteenth and sixteenth centuries the iron industry was so stimulated by its own improvements and by the advances in general civilization, and it made such serious inroads into the forests, that in 1558 the cutting of wood for iron making was prohibited in certain parts of Great Britain, and in 1584 the further building of iron works in Surrey, Kent, and Sussex was forbidden. The increasing scarcity of wood, due not only to the growth of this industry but to the advance of civilization in general and the increasing density of population, must have been among the chief causes of the attempts which were then made to replace charcoal with mineral fuel for iron smelting. The use of mineral coal for this purpose was patented by Simon Sturtevant in 1611; Dudley overcame the technical difficulties in the way of making both wrought and cast iron with this fuel in 1619, but unfortunately all his attempts to establish this industry were defeated by the opposition of the charcoal iron makers. Strada had no better commercial success in Hainault in 1625, and it was not till 1735, more than a century later, that Abraham Darby showed how to make cast iron by means of coke in what had by this time become a veritable blast furnace.

13. The Hot Blast.—In 1811 (note that the tempo was still an *ultra largo*) came the next great improvement, when Aubertot, in France, heated steel by burning the rich furnace gases, which till then had been allowed to waste at the blast-furnace top. Seventeen years later, in 1828, J. B. Neilson heated the blast, saving fuel to a degree not only wholly unexpected, but at first sight out of all proportion to this apparently simple step. Four years later (our *largo* has now given place to an *allegro*), in 1832, this heating of the blast was done by means of the furnace gases themselves, at Wasseralfingen.

14. Puddling.—Meanwhile, in 1784, Henry Cort had very greatly aided the conversion of cast into wrought iron by substituting for the old forge, in which the contact of the fuel with the metal made decarburization very difficult, the reverberatory "puddling" furnace, which burnt the fuel apart in one chamber, and in a separate one heated and melted the iron by means of the burning gases, the flame, which sprang from that fuels as sketched in Fig. 1.

During the latter part of this period the progress in mechanical engi-

neering stimulated the iron industry greatly. Thus in 1728 Payne and Hanbury gave us the sheet-iron rolling mill; in 1760 Smeaton's cylindrical cast-iron bellows began to replace the wooden and leather ones till then in use; and in 1783 Cort's grooved rolls for rolling rods and bars, and in 1838 James Nasmyth's steam hammer, very greatly cheapened the working up of the product. But a still greater stimulus was given between 1760 and 1770 by Watt's steam engine, and in 1825 by the railroad, each of which led to a great demand for iron for itself and for the industries which it served and stimulated. Moreover, each aided the iron master directly, the steam engine as a most useful tool, the railroad by cheaply assembling his materials and distributing his products.

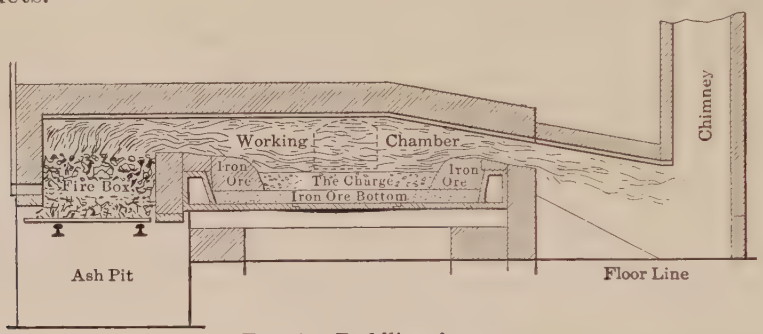

FIG. 1.—Puddling furnace.

15. Huntsman's "crucible process" invented about 1740,[1] of melting high-carbon steel in little crucibles in order to free it from the particles of iron silicate or slag which contaminated it, very greatly bettered the quality of steel, though the cost of thus melting these little lots is so great that the process has always been confined to the manufacture of steel for the most exacting purposes. Today its existence is seriously threatened by the electric melting furnaces.

16. The Bessemer Period.—The great achievement of this period is the development, at a moderate cost and on an industrial scale, of a temperature high enough to melt not only the most infusible varieties of steel but wrought iron itself, so bettering these products as Huntsman had bettered the relatively fusible high-carbon steel by cleansing them of their befouling slag. This was done first by Bessemer about 1855, and then by Siemens and Martin in 1864.[2] These inventions are of far greater importance than

[1] Indeed Réaumur had, in 1722, referred to experiments in which he had made steel by melting wrought iron and cast iron together in crucibles ("L'Art de convertir le fer forge en acier et l'art d' adoucir le fer fondu," 1722). The experiments do not seem to have led to any industrial result till after Huntsman had established the crucible steel industry. Long before Réaumur, the Hindoos had made Wootz, a high-carbon steel, out of crude iron sandwiched with wood in crucibles. This will be considered later. Here suffice it that the conditions of this operation, the extremely low density, the difficulty of forging, and the damaskeening, all go to support Ledebur's assertion that Wootz is a "weld steel," *i.e.*, that there is no complete fusion in this crucible treatment. (Ledebur, "Handbuch der Eisenhüttenkunde," 1908, vol. 3, pp. 273–4.)

[2] The important parts played by Mushet and Göransson in connection with Bessemer, and by Réaumur, Heath, and Louis Le Chatelier as forerunners of Martin, will be considered later. (See *Revue de Metallurgie, Memoires*, 1907, vol. 4, p. 212.)

Huntsman's, both because they work so very much more cheaply, and because the industrial usefulness of their products, which include rail steel and the numberless "structural" steels used for general engineering purposes, is incomparably wider than that of Huntsman's.

Looking back, we see that the distinctive work of both the cast-iron and the Bessemer periods is freeing the metal from enclosed slag by melting it. This was done in the cast-iron period for the more fusible varieties, cast iron and the high-carbon steels, and in the Bessemer period for the most infusible varieties, wrought iron and low-carbon steels.

Bessemer discovered both that the rapid oxidation and removal of the impurities of cast iron created a temperature high enough to keep molten the steel into which the cast iron itself was turned by that very removal, and the very great industrial value of the low-carbon steel which he was the first to make on an industrial scale. With this knowledge, and having in the Siemens regenerative gas furnace an independent tool for generating these very high temperatures, the brothers Martin, at Sireuil in France, worked out the "open-hearth" process of making steel of any needed carbon content, by melting together on the hearth of this furnace cast iron rich in carbon and scrap iron poor in that element. Their much less dramatic but much more plastic process seems sure to displace Bessemer's for most important purposes.

17. Thomas's Basic Process.—The great defect of both these processes in their original form, the acid Bessemer and Siemens-Martin or open-hearth, is that they do not remove the extremely harmful element, phosphorus, which contaminates nearly all ores of iron. This defect Thomas cured in 1878, when he proved publicly that, in the presence of a very basic molten slag, the whole of the phosphorus could be removed.

18. Metallography.—Next in value to the discovery of these processes comes the birth and rapid growth of the new science of metallography, which, together with Bell's studies of the blast-furnace process, have set the metallurgy of iron on a scientific basis. Sorby published important brief notes about the microscopic structure of iron in 1864, and Martens, quite independently, published notes on its microstructure in 1878; but the first very important step was taken by Tschernoff, who in 1868 enunciated the chief laws which govern the metallography of iron. These were supplemented in 1885 by Brinell's laws. In 1888, by a flash of genius, Osmond saw that the extraordinary influence which thermal treatment and the presence of certain foreign elements have on iron was due to allotropy. From this and later discoveries has come the remarkable development of the "special" or "alloy" steels which owe their properties to the presence of some element other than carbon. Prominent among these are the austenitic manganese steel, described in 1885 by Robert Hadfield, the first substance to combine marked hardness with great ductility, and the "red-hard" or "high-speed" tool steel of Taylor and White, which retains its hardness even when it is heated to dull redness by the friction caused by rapid cutting.

CHAPTER 2

THOUGHTS ON THE PERMANENCE OF OUR SUPPLY OF IRON[1]

19. Predictions of the Early Exhaustion of our Iron.—Are iron and steel about to go out of use, or is the impending exhaustion of our iron-ore supply about to oppress us severely by causing a sudden and enormous increase in the cost of iron? That some such calamity is close upon us might easily be inferred from much that has been written lately. For instance, if the consumption of iron were to increase hereafter as fast as it did between 1893 and 1906, the ten billion tons of ore, with which Professor Tornebohm credits the world in his report to the Swedish government, would be used up in about 40 years. This seems to mean that, in the lifetime of the young people of today, mankind will be confronted with the stupendous task of replacing iron with some other material, not only for machinery of all kinds, but also for such important objects as railroad rails and wheels, ships, high buildings, roof trusses, springs, cutting tools, and magnets.

We may almost say that the materials with which iron can be replaced are yet to be discovered. The other metals and their known alloys come nearest to being promising substitutes, but even apart from their far higher cost, they are much less fitted than iron for making most of these objects. What is far worse, the very exuberance of our activity, which at first sight seems to threaten us with an iron famine, is likely to exhaust the supply of the other metals at about the same time. At such a prospect we may well say with Burns—

> And forward, though I canna see,
> I guess, an' fear.

20. Reasons for Rejecting these Predictions.—Nevertheless, I believe that the cost of iron will not begin to oppress us within many hundreds, yes thousands, of years, if indeed it ever does.

How can this belief be justified? Not by pointing out how greatly Professor Tornebohm has underestimated our own iron-ore supply, which, according to Mr. Eckels of the United States Geological Survey, is at least ten, if not twenty, times as great as that with which the learned Swede credits us. Not by the consideration that only certain parts of Europe and a relatively small part of North America have thus far been explored carefully, and that the rest of these two continents, together with South America, Asia, Africa, and Australia, may reasonably be expected to have, collectively, enormous quantities of ore. Nor yet by pointing out that one of the rich-

[1] This chapter is reproduced, with slight changes, from an article by the Author, "An Optimist's View of the Iron Ore Supply," in the *Atlantic Monthly*, June, 1910, p. 827, with the kind permission of its publishers, The Atlantic Monthly Company, Boston.

est and most abundant of ores, pyrite, is not included in any of our current estimates. These considerations are important; they increase the expected life of our iron-ore supply from decades to centuries; but they are much less important than the fact that there is an incalculable quantity of material which, though not ore today, will become ore as soon as it is needed.

21. What is Iron Ore?—At any given time it is simply rock rich enough, in large enough masses, and near enough to the surface to be treated with profit in competition with the other iron-bearing rocks which man is then working. Rock with 2½ per cent. of gold is an extraordinarily rich gold ore; rock with 2½ per cent. of copper is copper ore today; rock with 2½ per cent. of iron is not iron ore today, for the sole reason that it cannot be worked at a profit in competition with existing richer rocks. It will become ore just as soon as the exhaustion of the richer rocks shall have enabled its owners to treat it with profit. Whether a given ferruginous rock is or is not ore, then, is purely a question of existing demand and supply. Most iron ores mined today contain at least 25 per cent. of iron, and some contain more than 60 per cent. As these richer ores are exhausted, poorer and poorer ones will come into use, until, to the eye of the prophet, a large bed of 4 per cent. ore, perhaps even of 2½ per cent. ore, becomes a veritable bonanza.

22. Other Substances will, like Iron, Increase in Cost.—This does not mean that iron will then be a semi-precious metal, because the cost of the other metals will rise like that of iron. It means only that iron may cost then as much as copper costs now. Our scale of cost will advance as a whole. Then we shall protect iron from theft, corrosion, and abrasion, as we now protect copper; and for that matter we shall then guard our copper as we now guard our gold, and watch our platinum as we now watch our diamonds. As the ranch is now, so will the four-acre lot be then.

Between these limits, the rich ores of today and the 2½ per cent. ores of that possible æon-distant time, there is an incalculably great quantity of potential ore, for as we descend in the scale of richness we ascend much faster in the scale of quantity stored for our use. How very rapid the ascent is may be inferred from the fact that the igneous rocks, which form a very large fraction of the entire crust of the earth, contain on an average about 4½ per cent. of iron, according to late estimates.

23. The Progressive Increase in the Working Depth of our Mines.—As with the workable richness, so with the workable depth. We think and speak today as if deposits of ore could in the nature of things be worked only to a very moderate depth, a few thousand feet. But this is an error. What is true is that the cost of working increases rapidly with the depth at which the work is carried on, so that at any given time the profitable depth of working is limited by the competition of ore from shallower mines. But, like the richness which makes an ore profitable, the profitable working depth is purely a question of demand and supply. The whole crust of the earth is ours. We will first take the richest ores, those in the largest masses, and

those nearest the surface, in short the most profitable ores; but we shall later take poorer and deeper ones. To this process there is hardly a limit.

Thus it is not a real iron famine that awaits us, but only the need of mining at greater depths and of handling more tons of ore and barren rock for each ton of metallic iron ready for man's use. This handling will in general have to include crushing the ore, and separating by mechanical process its scattered particles of minerals rich in iron from the great mass of barren minerals with which they are usually mixed.

24. The Iron Core of the Earth.—Here it may interest and perhaps profit us to speculate a little, adopting the plausible belief that the earth itself is a huge meteor, with a relatively thin crust of rocks, and attempting to confer on future generations, in our imagination, some fraction of the powers which they will surely develop. To any such speculation it is essential that we should remember the extreme crudity of our present civilization. Here we are but a few centuries past the beginning of the historic period, while before us lie untold millions on millions of years during which the world may remain inhabitable. Let us conceive that the Middle Ages, which in our vanity we put some centuries behind us, in fact lie millions of years before us; that we have hardly yet advanced well into the dawn of history; that the mechanical powers of our successors will exceed ours a thousand-fold more than ours exceed those of the cave-dwellers. This should be our attitude if we would have any approach to true perspective. We must remember that things which today seem impossible, are impossible only because of our present crudity and dense ignorance; that, if man's power over nature increases in the next million years at the rate at which it has increased in the few years since Watt gave us the steam engine, almost the only things mechanically impossible will be those which today are unthinkable and self-contradictory. From this point of view we can see the shaft of some bold syndicate piercing its way through the thin crust of rocks to the iron core. Impossible and inconceivable, most men will say! Inconceivable if we base our conceptions solely on our present development; conceivable if we take into account the probable development of man's mastery over nature. I do not attempt to spell out the exact mechanism of this exploitation. I simply say that, if man shall some day come to need that iron, he will, if he becomes the master that I picture him, make a way to use this mass which God has given him. I am looking forward, not hundreds, but thousands and millions of years.

25. The End of Mundane Life.—Once the core of the globe is reached, we shall have iron enough to last until the secular cooling of the globe shall have gradually crowded to nothing the area warm enough for vegetation, and thus shall have killed the last man by depriving him at once of his food and of his oxygen; or until our water or our oxygen has been dissipated into space, if it is possible for the kinetic force of these gases, in coöperation with the imaginable friction of the ether, thus to overcome gravity. The end of vegetation means not only starvation but suffocation, because it is

vegetation alone that regenerates the oxygen while we breathe. Long after we have begun to regain from the ocean the matter of which our present mountains are formed; beyond the vast ages in which we nestle ever deeper into the bosom of our great mother, where is stored for us that heat which, though squandered today, will then be as the breath of our nostrils; when the polar ice-caps slowly creeping outward at last meet at the equator; when Mother Earth, exhausted, draws together those icy curtains for her endless sleep, this vast store of iron will remain unlessened to mock the last of our race.

But may we not put off the day of death? May we not make up for the paling of our sunshine by devices for raising our food, and generating our oxygen, by means of the energy of waterfalls, wind, waves, tides, and the enormous momentum of the earth itself? Grant it: it is but a postponement; we should exhaust these sources of energy in turn; yet even then the store of iron would remain, because it is not destroyed by use but only dissolved or worn to powder, to reprecipitate and reconcentrate.

26. Utilization of the Ether.—We often speak as if the momentum of the earth could not be utilized by those standing upon it, forgetful that the ether through which it is plunging is a fulcrum—not the most convenient one possible, but yet a fulcrum—whence it may be pried. A man on a cannon ball rising through the air might utilize the momentum of the ball itself, by means of a windmill driven by the friction of the air through which he passed. Thus man, living on our earth as it rushes through the ether, may some day learn how to utilize the earth's momentum by means of an ether mill, a mechanism driven by the friction of the ether through which he is moving. In doing this he would retard the earth's speed, be it ever so gradually, and cause it the sooner to run down, and come to rest upon the sun. Indeed, as the sun cools, we may purposely delay the earth's cooling, and thus prolong the age through which it will remain warm enough to support life, by retarding its velocity, and thus bringing it nearer to its source of heat. Thus may man one day modify the climate of our planet. Who shall say that he may not in time modify the seasons themselves—improving nature's scheme in this grand way?

We know too little about the properties of the ether to speculate to advantage about the mechanism of an ether-driven mill. But however strikingly it may differ from all other forms of matter, it is after all only matter, to be harnessed in due time by the race which has in its infancy already learned to speak by the lightning.

27. The Iron Core Inexhaustible.—If we form a picture of man's beginning his drafts on this iron core; of his putting successive lots of this iron into use; and of the gradual rusting and wearing to powder of lot after lot, till those drafts shall have amounted to any considerable fraction of the core itself; if we remember that this iron rust and powder will be spread out over the face of the earth, or at most be carried by solution into the earth's crust; and if we bear in mind that this crust is probably but a few hundred

miles deep, whereas the iron core is thousands of miles thick; we find it hard to escape the conviction that this vast quantity of iron rust and powder must in itself form reconcentrations upon which man may draw for his use. Indeed, if all this iron were distributed evenly throughout the earth's crust, that whole crust would thereby become a tolerably rich iron ore. The essential thing to recognize here is that, although iron disappears from sight as dust or rust, it is not destroyed, but is ever accumulating. Thus the supply of iron is not simply relatively, but absolutely, inexhaustible. From this point of view it is indeed possible that, even before we shall have used up the accessible iron in the earth's crust, these reconcentrations will form rapidly enough to supply with iron a population greater than that which the growth of vegetation can support.

Argument is hardly needed to show how impossible it would be for man to put into simultaneous use any large fraction of the huge mass of iron contained in the earth's core. As we attempt to conceive our successive drafts on this vast fund, looking at them first from one standpoint and then from another, we see limiting conditions which would arrest us before we had drawn and used any large fraction of the supply stored to our credit. For instance, if we consider the conversion of iron into frame buildings, set side by side in actual contact one with another all over the earth, and if we imagine that some marvelous engineer, some giant, outstripping Napoleon in genius ten times as far as he outstripped the commonest dullard, could raise these buildings untold miles above the level at which his workmen could breathe; before even such an enterprise could use up any significant fraction of the iron of the earth, those buildings would stretch far above the upper surge of all, the majestic billows of thinnest air which may tower and rage at the surface of our atmosphere.

Then, again, before any large fraction of our iron could be put into use, the incidental rusting of successive masses of it which meanwhile had done their work and disappeared from sight would so far rob the atmosphere of its oxygen as to make it unbreathable.

Pondering thus, the Martian would smile incredulously if told that man, living as he does on an iron-cored globe, was fretting about the exhaustion of the most exhaustless of his supplies.

Oxygen and food can be had only so long as energy remains. The store of terrestrial and solar energy is finite; its passing is but a day in the life of the universe, and when it is exhausted life must cease. But the earth cannot get rid of the iron of which so great a fraction of its very self is formed.

28. What Hardship will the Increase in the Cost of Iron Cause in the Centuries Near at Hand?—It is of more immediate interest to consider whether man is likely to be oppressed by the increased cost of iron in the early centuries of the period between the present and that immeasurably, yes inconceivably, distant day when we shall gain access to the earth's central mass of iron. Two questions suggest themselves:

First, how rapidly will the increase in the cost of ore raise the cost of the iron made from that ore?

Second, will this prick be so much sharper than the others which must accompany it that it will be felt as a hardship different in kind from the rest?

As regards the first question, let us remember that the cost of iron in the ore is only a fraction of the cost of the finished iron articles themselves, hardly a quarter of the cost of even such crude products as rails, and an insignificant part of the cost of many of the important finished articles, such as springs, cutting tools, wire, etc., now made of iron, and less readily made from animal or vegetable matter. If, through the necessity of using ores much poorer than our present ones, the cost of a ton of iron in the ore, concentrated and ready for smelting, were to double, the cost of a ton of rails would increase by only one-quarter, and the cost of a table knife by a trifling fraction. Indeed, by increasing the cost of the fuel needed for treating our ores, the rapid exhaustion of our coal-fields may do more than the growing poverty of those ores themselves to increase the cost of iron.

29. Will the Rise in Wages Raise the Cost of Iron?—You may ask whether the cost of iron to each of us is going to be raised seriously by the continuous and irresistible rise in wages. Probably not. Given a world of men nearly all of whom work in one way or another, the chief permanent element of cost of each product to each consumer is the number of days of labor and of care which its production requires. Note clearly that this is the true measure for our present purpose. The fact that our future ores will be poorer and deeper seated than our present ones may raise the cost of iron thus measured, but the rise of wages should not, as we can readily see on reflection.

Measured as they should be by their purchasing power, wages have risen with the secular rise in the scale of living which has given the almshouse pauper what would have been fabulous luxury to Siegfried, Achilles, or Noah. But though this rise should certainly continue, it does not imply an increase in the cost of iron or other products, measured by the true scale—days of labor. This increase in the quantity of goods which society, or mankind collectively, returns to each worker for each day's work is of course possible only because of the greater quantity of goods which, thanks to improvement in the processes of manufacture and distribution, society gets from each day's labor. What all men collectively can distribute to all individual workers severally is nothing more than the sum of what all men collectively have received from those individual workers, be they laborers, superintendents, presidents, or financiers. Practically speaking, all but an insignificant fraction of us are workers in one way or another. If we each get more for our sweat, or for our thought, than our peers did in Noah's day, it is possible only because as a whole we create more with a day's work. The increased purchasing power of a day's wages simply reflects the increased producing power of a day's labor. A hundred years hence society, as repre-

sented by the shopkeepers, will give the workman who makes iron more goods in return for a day's wages than it gives him today. But with the labor which he gives for those wages he will then make proportionally more iron than now for society, as represented immediately by his employer, and ultimately by you and me, his employer's ultimate customers; so that society, you and I, will give him no more goods for his pound of iron than we give him today. He will get, and we shall give, more goods for a day's labor, but not more for each pound of iron that he makes.

30. The Cost of Iron will Rise more Rapidly than that of Vegetable Products.—Turning to our second question, the cost of iron and of the other metals, indeed of all mineral substances, seems likely to increase relatively to the cost of food, wool, cotton, leather, and other vegetable and animal products, because, though our present rich and superficial ore bodies are not re-creatable, and though we must turn ever to poorer and deeper-seated ones, the supply of vegetation seems to be not only re-creatable at will, but susceptible of enormous increase. I do not here consider the very distant time when the natural re-concentration of the metals worn out and rusted away in use shall have begun to form useful ore deposits. We may assume that we shall soon cease to squander our little store of phosphates, potash, and other mineral matter necessary for plant life by washing them away with the waste products of city life, out through the sewers into the rivers, and thence to the ocean; and that, within a measurable time, we shall treat all sewage so as to recover these mineral matters, and return them to the land. This we shall certainly do, for if we do not we die. These precious substances are no more re-creatable than iron ore, but fortunately they are recoverable by simple means.

Yet even so, one cannot feel quite sure that vegetation will continue indefinitely to be abundant. It is possible that the cultivation of the land by civilization necessarily implies that the spring floods must, century by century, wash away into the ocean the thin layer of plant-sustaining top soil which long ages have accumulated, and wash it away faster than it can be regenerated. Even if we impound the precious mud which is now heedlessly wasted, and dredge or hold it back from every ocean, even then will it not slip from us faster than it can be renewed by the decay of rocks? This is conceivable.

31. Will Iron Rise in Cost Faster than Other Inorganic Products?—But will the increase in the cost of iron ore be felt more sharply than that of other mineral matter—gold, copper, and the other metals, coal, lime, cement, and the precious stones?

It will be felt more than the increase in the cost of most other metals, not through its being more rapid, but because we are more dependent on iron than on any other metal, indeed probably more than on all the others put together.

The increase in the cost of iron will be quicker and will be felt more sharply than in that of lime and cement, because the stores of limestone and cement-

yielding rock are so vast. In the same way, it may be much quicker than the increase in the cost of aluminum, because of the great quantity of this latter metal which our clays and crystalline rocks contain.

As the cost of iron increases relatively to the cost of the vegetable products, or of those mineral products of which such vast stores are at hand, and as this relative increase begins to be felt, it will retard itself by checking our present waste of iron, and by checking also the demand for iron, which will be replaced by those other and cheaper substances. Indeed, concrete has already replaced iron for certain purposes, because it has become so cheap.

32. The Appreciation of Iron less Serious than that of Coal.—But the increase in the cost of iron will be only a pin-prick beside the loss of our coal; first because coal is much more important than iron—indeed the money value of the coal which we use is five times that of the iron ore; next because, in the nature of the case, dearness of coal, which is our fuel for heating and cooking, pinches the individual much more sharply than dearness of iron can; and finally because coal will advance in cost incomparably faster than iron, indeed at an ever-accelerating rate.

For this there are two chief reasons. First, the fact that most coal exists in the carboniferous formation has led to the systematic exploration of that formation in many countries, so that we already have a rough knowledge of how much coal there is in Europe and North America, though we cannot even guess at the quantity of iron ore, because iron ore may exist in any geological formation.

Second, because it is only in the relatively thin layers of rocks which represent the short period since vegetable life began on the globe that we can hope to find coal, while iron we may confidently expect to discover throughout the enormously greater mass of rocks underlying them, and reaching the iron core itself. Indeed, the deeper we go the richer should those rocks be in so heavy a substance as iron.

Of coal it is a true famine that lies ahead; of iron, it is merely a gradual increase in cost; for, whereas there is only a trifling stock of coal, and this is absolutely destroyed in use, the exhaustless stores of the indestructible metal of Mars are even now renewing themselves, so that before the immeasurably distant future when any large fraction of our iron shall have been in service, new masses of it will have re-concentrated ready to our hands. Then iron will indeed have grown dearer, like horses, cotton, Greek vases, and land; but coal will have vanished forever, like the ichthyosaurus, to be replaced by energy from waves, wind, sun, and ether.

33. Summary.—Let me sum up my argument. When we come to consider the effect which the drafts on our iron-ore resources will have in increasing the cost of iron as measured in days of labor, we see that this effect will not be felt oppressively until after an incalculable age, if indeed it ever is: first and chiefly because of the inconceivable immensity of the stores of potential poor and deep-seated ores; second because the cost of the iron in the ore is only a fraction, often only an insignificant fraction, of the

cost of the iron articles of our actual use, so that a considerable percentage of increase in the cost of the former causes only an inconsiderable increase in the cost of the latter; and third because our sense of this increase will be dulled by the simultaneous rise in the cost of almost all other tangible things. To the hardship of this general rise, even to the total disappearance of that most precious deodand, our coal, we shall certainly learn to accommodate ourselves. And though the fraction of this hardship which consists of the increase in the cost of iron may be greater than some of the other fractions of discomfort, such as those formed by the advance in the cost of the other metals, and of certain animal and vegetable products, yet it will not differ from the accompanying fractions taken severally, and certainly not from their sum, in such a way as to be felt oppressively and apart from the rest.

Let us husband our iron ore, like every other gift of God. But when we cry out to others to join in the husbandry, let our appeal square with the facts, that those who hear may heed.

CHAPTER 3

OUTLINE OF THE CLASSIFICATION AND MANUFACTURE OF IRON AND STEEL

34. What are the Chief Kinds of Iron and Steel?—Iron differs from most of the other metals in not being made industrially in the pure state, but in the form of an alloy, both because of the difficulty in removing the last small quantity of certain elements, especially carbon, taken up in extracting the iron from its ores, and because pure iron is too soft and weak for most uses. Hence while the name "iron" in one sense denotes the element itself, quite as "gold" and "copper" do, in the usual collective and industrial sense it denotes a great group of alloys in which iron is the dominant element. They are strictly iron alloys, but we call them for convenience "iron."

Among these industrial alloys which make up the "iron" of commerce the carbon-iron alloys far outweigh the others in importance, essentially because they are much cheaper than the others. This in turn is because carbon is so much cheaper than any of the other alloying elements, and indeed because it is unavoidably present in the iron as it is extracted from its ores in contact with carbon in the iron blast furnace.

As is the case with so many other metals, iron is strengthened but embrittled by most of these alloying elements, and especially by carbon. Hence the carbon-iron alloys are divided into those with so little carbon that they are usefully malleable, and the cast irons which contain so much carbon that they are not.

These latter owe their use to their being cheap, both because it is in this greatly carburized state that the crude or cast iron is made in the iron blast furnace, and because this cast iron is so fusible that it is readily cast directly into the shapes in which it is to be used in the arts.

The malleable iron-carbon alloys, steel, wrought iron, and malleable castings, are actually derived from cast iron (see §37), either by changing its carbon to a relatively harmless state so that in spite of it the iron is usefully malleable, as in the case of "malleable castings," or by reducing that carbon to a proportion so small that in spite of it the metal is usefully malleable, as in the case of steel and wrought iron.

Of this iron thus made malleable by removal of carbon, one particular species is set apart under the name "wrought iron" or, in Great Britain, "malleable iron," and all the others are called "steel" in English. Wrought iron has two distinguishing and essential though unrelated characteristics: (1) That it contains so little carbon that it is relatively soft and ductile even after a quenching in water such as makes the higher carbon steels extremely hard and brittle, and (2) that it is made by welding up pasty white hot particles

of iron in a bath of molten iron silicate or "cinder," §10, with the result that much of this cinder remains mechanically imprisoned in the iron. The remaining kinds of iron made malleable by the removal of carbon are called "steel" in English, whether they lack either or both these essentials of wrought iron, so that some steel is steel solely because it has more carbon than wrought iron, and some is steel because it is made without this welding up in a bath of cinder, in short because it is nearly free from cinder. The former class, consisting chiefly of blister steel, is relatively unimportant, so that the great bulk of the steels of today are simply the malleable kinds of iron which have been cast from a molten state instead of being aggregated like wrought iron from pasty particles in a bath of cinder. In fact blister steel is made by the roundabout process of first decarburizing cast iron and thus converting it into wrought iron, and then again carburizing that wrought iron by long heating in contact with charcoal. Because it retains the cinder which became imprisoned in it at the time when, as a plastic or pasty mass, its wrought-iron particles were welded together, it is called "steel of plastic origin" to distinguish it from the remaining kinds called in English "steel," which are freed from cinder by casting them while molten into a malleable mass, and hence are called "steels of molten origin."

In addition to the carbon-iron alloys there are the so-called "special alloys," the special steels and the ferro-alloys.

The special steels, called also the "alloy steels," differ from the normal or carbon steels in owing their properties primarily to the presence of some element other than carbon. Of these elements the chief are chromium, manganese, nickel, and tungsten. But these special steels, though not unimportant, form only a minute fraction of the whole.

The ferro-alloys differ from the special steels in having so large a proportion of the alloying element that they are used as a vehicle for introducing this element in the course of manufacture. Ferromanganese, for instance, is an alloy usually of about 80 per cent. of manganese, 15 per cent. of iron, and 5 per cent. of carbon. Thus strictly speaking it is rather a manganese alloy than one of iron, though classed with the ferro-alloys because it shares their function. An alloy, such as that of 25 per cent. of nickel and about 75 per cent. of iron, may be both a special steel and a ferro-alloy, fulfilling the functions of both.

The great bulk of the iron of commerce consists of cast iron and the "normal" or "carbon steels" of molten origin, which latter form the chief subject of this work, though the cast irons which are separated from the steels by an arbitrary line, the position of which is in dispute, must needs receive a large share of attention in the broad survey which the present volume attempts.

These normal or carbon steels and cast iron form a continuous series. They consist essentially of iron and carbon. The properties of any given specimen are due primarily to its carbon-content, and in the case of cast iron to the condition of that carbon, though they may be modified greatly by

thermal and mechanical treatment, and also very materially by the effects of a little manganese and silicon intentionally added, and of a little phosphorus and sulphur, two harmful elements whose presence is tolerated. Small quantities of oxygen, hydrogen, and nitrogen also have their effect.

35. The General Scheme of Iron Manufacture. The Conversion of the Iron of the Ore into Cast Iron.[1]—The ultimate source of all kinds of industrial iron, both malleable and unmalleable, is the iron ore of the earth's crust, chiefly oxides and carbonates.[2] The first step in the manufacture of the several varieties of iron and steel, including the unmalleable cast iron and all the malleable varieties,[3] is to bring the iron of the ore to the condition of cast iron, which in turn may be converted into one or another of the malleable species. This indirect procedure in making these malleable species is not indeed an absolute necessity, but it is more economical than making them direct from the ore, for reasons which we shall see beyond.

The conversion of the iron of the ore into cast iron is brought about in a huge vertical shaft, the iron blast furnace, Figs. 2 and 3, filled from top to bottom by a column composed of lumps of ore, of coke, and of limestone, which are charged at short intervals at its top. This vast column sinks slowly as the heat generated by the combustion of its coke with pre-heated air blown in at *O* melts its base away; and as it thus sinks, its top is continuously replenished by additions of lumps of ore, of coke, and of limestone.

The combustion of the coke forms carbonic oxide gas, a very powerful deoxidizing agent, which, as it rushes swiftly up through the overlying column,

[1] I hope to treat these industrial processes at great length in a later volume. The following works may be consulted:

Campbell, H. H., "The Manufacture and Properties of Iron and Steel," McGraw-Hill Book Co., Inc., New York, Fourth edition, 1907.

Carnegie, assisted by Gladwyn, "Liquid Steel," Longmans, Green & Co., New York, 1913.

Dichmann, "Der basische Herdofenprozess," Julius Springer, Berlin, 1910.

Idem, "The Basic Open Hearth Steel Process," translated by Reynolds; Constable & Co., London, 1911.

Giolitti, "The Cementation of Iron and Steel," translated by Richards and Rouiller, McGraw-Hill Book Co., Inc., New York, 1915.

Hall, "The Steel Foundry," McGraw-Hill Book Co., Inc., New York, 1914.

Harbord and Hall, "The Metallurgy of Steel," Griffin & Co., London, Fourth edition, 1911.

Sexton and Primrose, "Metallurgy of Iron and Steel," Scientific Publishing Co., Manchester, Second edition.

Stoughton, "The Metallurgy of Iron and Steel," McGraw-Hill Book Co., Inc., New York, Second edition, 1911.

"Verein deutscher Eisenhüttenleute, Düsseldorf,"1912, *Gemeinfassliche Darstellung des Eisenhüttenwesens.*

West, "Metallurgy of Cast Iron," Penton Publishing Co., Cleveland, O., Fourteenth edition, 1912.

[2] Pyrites, FeS_2, also supplies a small quantity of iron, and with the exhaustion of the most easily worked oxide beds will probably supply much more. But a first step in its treatment is to convert it into oxide by removing, and usually recovering, its sulphur; so that broadly speaking, as the ores of iron reach the iron metallurgist they are oxides or carbonates.

[3] A very little wrought iron and steel is made directly from the ore by certain crude processes; and, in the open-hearth process of converting cast iron into steel, the iron of a very small quantity of ore is brought directly to the state of steel without passing through that of cast iron.

accompanied by much atmospheric nitrogen, deoxidizes the iron oxide of the ore, bringing the iron to the metallic state. Thus the iron, with the gradual descent of the column, is first deoxidized, then further heated by soaking up the heat which the rising gases bring up from the region of combustion, and then melted at *A* as it nears this region. Then it trickles down and collects at the very bottom, where it is overlain by a layer of molten slag, a silicate formed by the union of the barren minerals of the ore with the limestone which was charged with the ore for that purpose, and accompanies it in its descent and fusion. Thence the molten iron and molten slag are drawn at intervals by opening the holes *L* and *P*.

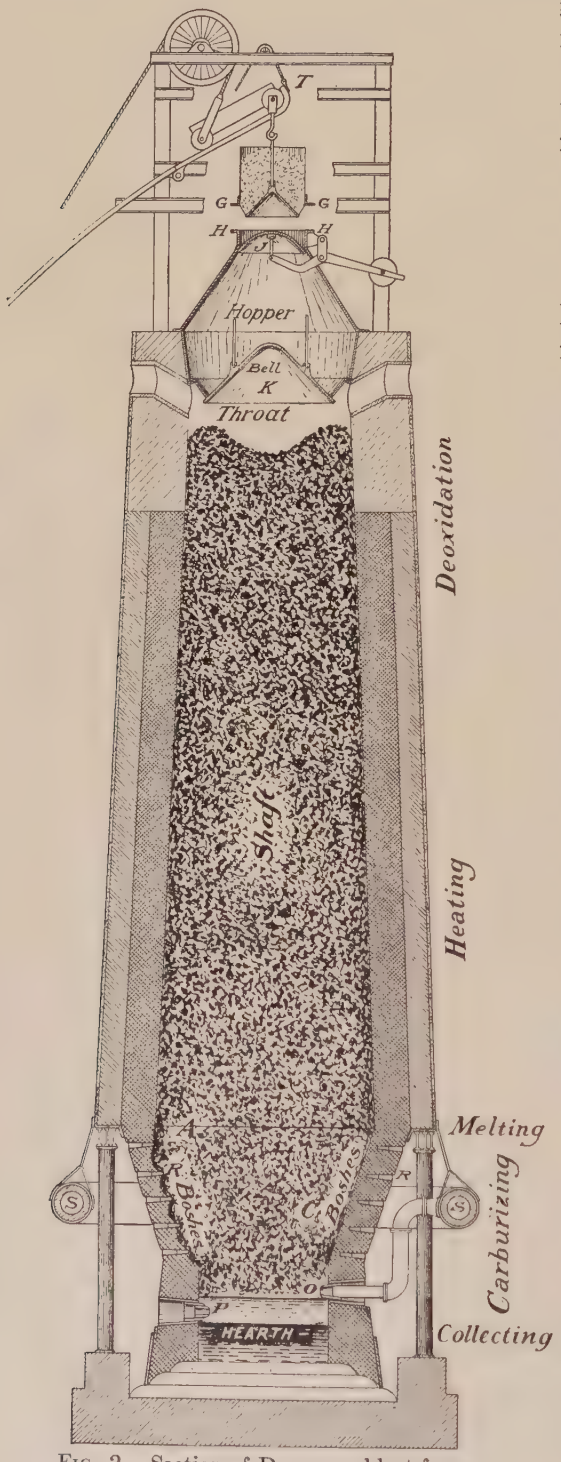

FIG. 2.—Section of Duquesne blast furnace.

The essential reason why this indirect method is the cheaper is that it removes cheaply and effectively the oxygen of the ore, and the sulphur and the barren minerals with which the ore itself, as it occurs in the earth's crust, is mechanically mixed. The removal of the oxygen is thorough, because of the strength of the deoxidizing conditions when the molten charge at last trickles down over the column of white-hot coke, and lies at the bottom with that coke dipping into it. It is effective, because the iron incidentally takes up so much carbon that it does not re-oxidize and hence re-unite with the barren minerals on its removal from the furnace, as it would if it were deoxidized without being carburized, and were then with-

drawn in the solid state. The removal of the barren minerals by this fusion is thorough; and a very large proportion of the sulphur is removed, because the excess of lime and of carbon, and the high temperature at the bottom of the furnace, collectively favor the desulphurizing reaction:

$$FeS + CaO + C = CO + Fe + CaS. \qquad (1)$$

The resultant calcium sulphide is nearly insoluble in the molten iron, and therefore passes out of the iron into the overlying slag, though the ferrous sulphide which it replaces is readily soluble in the iron.

The reason why the process is economical is (1) that it utilizes with extraordinary completeness the calorific power of a cheap fuel, coke, thanks to the thorough transfer of heat from the gaseous products of the combustion of that fuel to the charge, with which their contact in transit is so extended; (2) that the carburizing of the iron makes it so fusible that it is brought cheaply to the molten state indispensable to separating it from the barren minerals, and is maintained readily in that molten state in which it is cast or carried to the steel-conversion apparatus far more cheaply than if it were solid; and (3) that the blast-furnace process, though complex in its essence, is so extremely simple in its outward form that it lends itself readily to working on an enormous scale, with extremely small cost for labor.

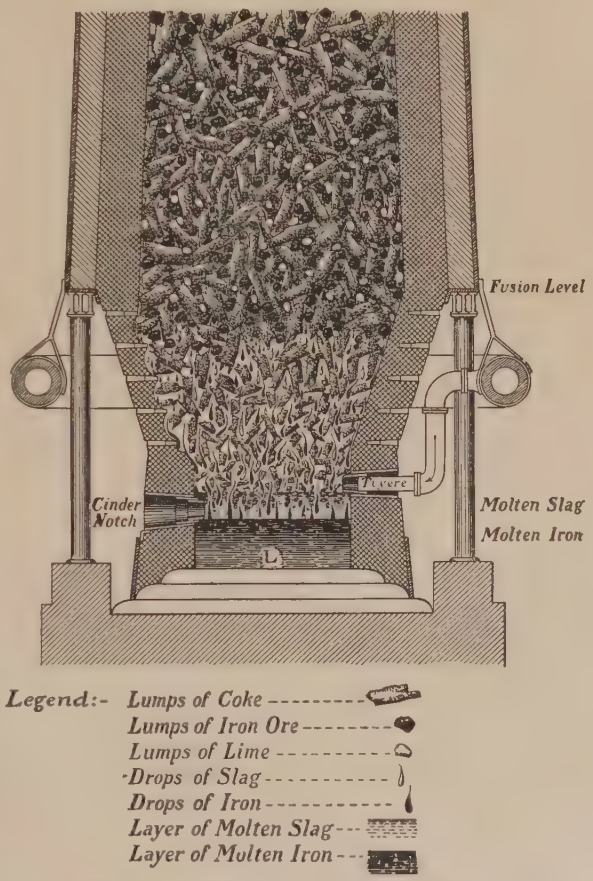

FIG. 3.—Lower part of the blast furnace.

In a word this indirect way of first carburizing and then decarburizing is expedient (A) because carburized iron is fusible and hence is melted cheaply, and so (1) is prevented from re-oxidizing, (2) is readily parted from the barren minerals, and (3) readily moved from place to place; (B) because the strong carburizing conditions insure the thorough deoxidation and recovery of the iron and the removal of much of the sulphur; and (C) because the deoxidizing, carburizing and melting machine, the blast furnace, utilizes the heat of a cheap fuel very thoroughly.

36. Difficulties in the Way of a Direct Process for Making Steel.—The first is that of so controlling the process as to deoxidize the iron and thus extract it from the ore, without carburizing it. This difficulty arises from the fact that at present the only cheap deoxidizing agents are carbon itself and carbonic oxide, and the latter, under conditions very likely to arise in practice, deposits carbon rapidly within the ore by the reaction:

$$2CO = C + CO_2. \qquad (2)$$

Of these agents, carbonic oxide indeed removes the first part of the oxygen of the ore rapidly and cheaply, partly by means of the carbon which it deposits within the ore itself; but it is not strong enough to be a convenient agent for removing the last of that oxygen. To do this readily needs actual contact with carbon. That carbonic oxide should have but little power to remove the last of the oxygen is readily understood when we learn that the molten iron may retain as much as 0.05 per cent. of oxygen even when it contains dissolved in itself as much as 4.06 per cent. of carbon.[1] If even carbon has such limited powers of extracting oxygen, what can we expect of its oxide?

Thus to deoxidize iron ore thoroughly without carburizing the resultant metal iron is very difficult if carbonic oxide alone is used, and of surpassing difficulty, if not indeed industrially impossible, if carbon itself is used, because the use of carbon implies actual contact with carbon, which the iron itself absorbs greedily.

If we go a step farther and consider the two possible paths of a process for making steel directly from the ore, (1) deoxidation without carburization and without fusion and (2) deoxidation without carburization but with fusion, we run into further difficulties. To deoxidize without fusion leaves the barren minerals and the sulphur of the ore to be removed by a later process. But we can hardly hope that any other process will remove either of these as cheaply as the blast furnace does. Further, it leaves the resultant iron in a spongy state offering such enormous surface that, if exposed to the air while hot, it re-oxidizes very rapidly; whereas to cool it before that exposure not only wastes its sensible heat, but implies troublesome and probably costly precautions to prevent re-oxidization in whatever process is then used for converting the spongy iron into steel.

To deoxidize with fusion and without carburizing is very difficult, for if there is contact with carbon the metal carburizes fast, and if there is not the task of making the deoxidation thorough is very difficult. An imperfect deoxidation implies not only loss of iron, for any undeoxidized iron not only remains mixed with the slag and so is lost, but makes that

[1] J. E. Johnson, Jr., private communication, Feb. 9, 1914. In addition to this carbon this iron contains 2.56 per cent. of silicon and 0.76 per cent. of manganese. Mr. Johnson finds that the presence of oxygen leads to the assembling of the graphite in flakes much smaller than those which form in its absence, a result to which we may refer the great strengthening effect of this oxygen in his tests. In my private practice I have found that this strengthening effect of a relatively large oxygen-content has been observed elsewhere independently. See §300, Chapter 11.

slag so corrosive that special and costly steps must be taken to prevent its destroying the walls of the furnace.

Under special conditions it may be possible for electro-thermal processes to reduce iron direct from the ore in competition with the blast furnace; but, till the ratio of the cost of coke energy to that of electrical energy becomes much greater than it is now, these conditions are likely to be extremely rare.

37. The further steps in the manufacture of iron and steel are shown graphically in Fig. 4. Starting with the conversion of the iron of the ore into cast iron in the iron blast furnace as roughly outlined in §35, the resultant slag,

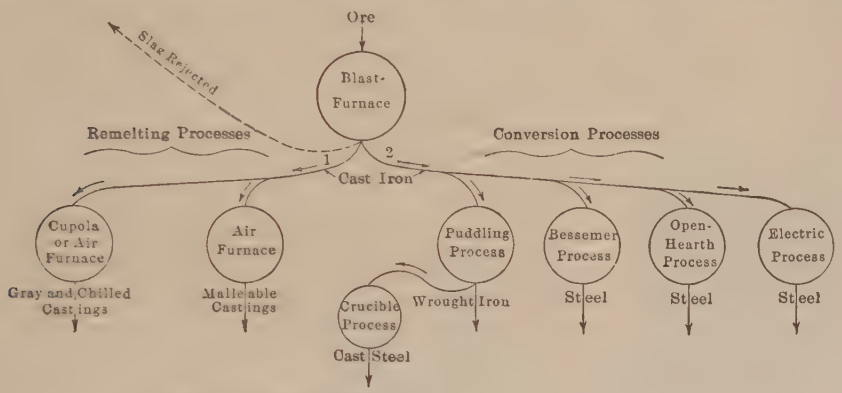

Fig. 4.—The general scheme of iron manufacture.

which contains the relics of the barrren minerals of the ore, the lime of the limestone, and the ash of the coke, is rejected, or used for filling in embankments, for railroad ballast, for making cement, and to a small extent for making slag wool, etc.

38. Iron Foundry Processes.—The cast iron may, following path 1, be cast, while it still remains cast iron, into castings which are to be used as such in the arts; or, following path 2, it may be converted into wrought iron or steel, which may be used in the common form of rolled or hammered products, or in case of steel in the form of steel castings.

If path 1 is followed the castings may be either (A) gray, (B) chilled, or (C) "malleable." The quantity of gray cast-iron castings used is incomparably greater than that of either chilled or malleable castings.

(A) In making gray cast-iron castings, the crude cast iron undergoes very little chemical change, indeed simply that incident to its re-melting in the foundry for the purpose of casting it, usually with some removal or dilution of its carbon, and such an adjustment of its silicon content as will cause a suitable proportion of its carbon to remain in the "combined" state as cementite instead of free as graphite. This re-melting is usually done in the cupola furnace, Fig. 5, a vertical shaft like the iron blast furnace but on a much smaller scale. Pigs of iron with lumps of coke, and often some flux to form a fusible silicate or slag with the ash of the coke, are charged

through the doors at the top, thus forming a column of solid materials which is continually replenished at the top as its bottom is burnt and melted away by the air blown in through the tuyères and the heat generated there by the combustion of the coke. The molten iron collects at the bottom, and the slag floats on its surface. Each of these is drawn off from time to time through the spouts shown. At the end of the day's work the furnace is emptied by dropping the doors *D*.

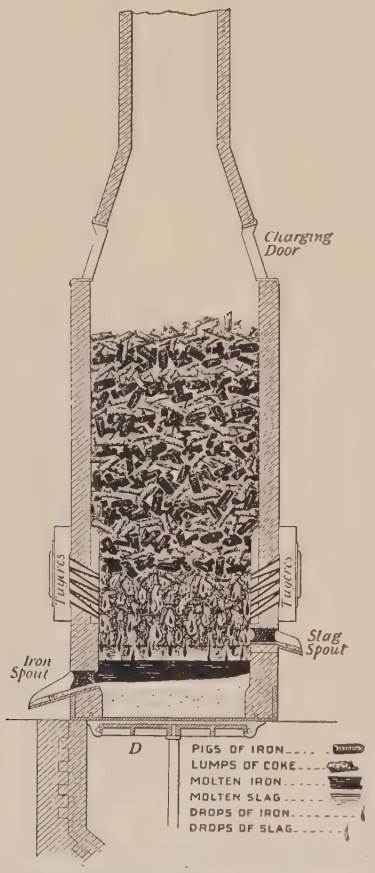

FIG. 5.—Cupola furnace for remelting pig iron.

(B) In making chilled cast-iron castings, though the ultimate composition of the cast iron is changed only in this same limited way, special hardness is given by chilling, *i.e.*, cooling the metal rapidly when it is poured into its moulds. In many cases further metamorphic changes, *i.e.*, changes in the microscopic constitution, are brought about by heat treatment (annealing).

39. In making malleable castings the ultimate composition of the metal is first changed somewhat by an oxidizing melting, which, however, leaves the metal still in the condition of white cast iron, an exceedingly brittle but fusible substance containing about 2.75 per cent. of carbon. This is cast into the various rather thin forms in which the metal is to be used in the arts, and these brittle castings are then made relatively strong, and relatively malleable in the cold, by a long heating or "annealing" process, which acts chiefly through inducing a metamorphic change, *i.e.*, a change in the microscopic constitution, changing the condition of the carbon and iron from that of the glass-hard, brittle cementite, Fe_3C, to that of soft ductile free iron or ferrite, intermingled with very finely divided graphite. Moreover, a small quantity of the carbon is removed by surface oxidation.[1]

These castings are made initially out of very brittle white cast iron, because this is so fusible that it can be cast cheaply even into very thin shapes. They are then made malleable either by removing part of their carbon by long heating (annealing) while packed in iron oxide, for instance

$$3Fe_2O_3 + C = 2Fe_3O_4 + CO, \qquad (3)$$

or by converting this carbon, by a like annealing, from its initial state of ce-

[1] *Cf.* Moldenke, "The Production of Malleable Castings," Penton Publishing Co., Cleveland, O., 1910. The percentage of carbon removed is about 0.45 per cent. in castings ½ in. thick.

mentite, the hardness and brittleness of which gives the initial white cast iron its brittleness, to the state of very finely divided or "temper" graphite,

$$Fe_3C = 3Fe + Gr. \quad (4)$$

In any individual annealing both these reactions take place, but in European practice it is chiefly the former and in American chiefly the latter that is relied on. The product of the European or decarburizing process amounts to some 50,000 tons a year, and is known as "white heart" or steely malleable. That of the American practice amounts to some 800,000 tons a year, and is called "black heart," because the presence of the finely divided temper graphite gives it a black fracture. Black heart annealing is for about 60 hours at about 730°C., white heart annealing for about 110 hours at between 800° and 900° or 950°, and it is said even 1,000°C.

40. Conversion or Purification Processes for making Steel and Wrought Iron.—If path 2 is followed, the metal has to undergo a very great purification, and it is of this purification that its conversion into wrought iron or steel really consists.

To explain, the blast-furnace process by which the cast iron is made is necessarily a strongly carburizing one, so that the cast iron necessarily contains much carbon. Further, the iron ore usually contains much silica (SiO_2), and some phosphoric acid in the form of apatite ($3CaO, P_2O_5$), together with more or less manganese in the form of manganese dioxide (MnO_2) and sulphur in the form of pyrites (FeS_2). The blast-furnace process is so very strongly deoxidizing that most of this phosphoric acid is deoxidized, as is much of the manganese dioxide and some of the silica; and the unoxidized phosphorus, manganese, and silicon which result unite with the molten cast iron, because they are unoxidized. At the same time some of the sulphur initially present as pyrites passes into the cast iron, which thus consists of metallic iron contaminated with these other elements, or impurities.

The following reactions may serve as types of those by which the deoxidation takes place in the blast furnace:

$$SiO_2 + 2C = Si + 2CO \quad (5)$$
$$P_2O_5 + 5C = 2P + 5CO \quad (6)$$
$$MnO_2 + 2C = Mn + 2CO \quad (7)$$
$$Fe_2O_3 + 3CO = 2Fe + 3CO_2 \quad (8)$$
$$FeO + C = Fe + CO \quad (9)$$

The essential difference between cast iron on one hand and wrought iron and steel on the other, is that the former always contains much more carbon, usually more silicon, and often more manganese, phosphorus, and sulphur than are permissible in the latter; and the essence of all the processes by which cast iron is converted into wrought iron or steel is the elimination of these foreign elements. The difference between the two classes may be illustrated by the following cases:

	Cast iron for the basic Bessemer process	Basic Bessemer steel, made from this cast iron	Difference removed in conversion
Carbon, per cent	3.50	0.10	3.40
Silicon, per cent	1.00	0.01	0.99
Manganese, per cent	1.80	0.50	1.30
Phosphorus, per cent	1.80	0.07	1.73
Sulphur, per cent	0.10	0.07	0.03

41. Chemistry of Purification.—The removal of carbon, silicon, phosphorus, and manganese is brought about by oxidation, and so is part of the removal of sulphur. The ultimate source of the oxygen may be either the atmospheric air as in the Bessemer process, or iron oxide such as native magnetite as in the puddling process, or both these jointly as in the open-hearth process. But even when atmospheric oxygen is used it appears to act indirectly rather than directly. That is to say, the atmospheric oxygen appears to act by oxidizing some of the iron itself to ferrous oxide, FeO, or by oxidizing ferrous oxide to magnetic oxide (FeO to Fe_3O_4); and the actual oxidation of the carbon and other foreign elements seems to be effected chiefly by the means of the iron oxides thus formed instead of by means of the atmospheric oxygen directly.

The reactions by which the oxidation takes place are of the following types:

$$2P + 5Fe_3O_4 = P_2O_5 + 15FeO \quad (10)$$
$$Si + 2FeO = SiO_2 + 2Fe \quad (11)$$
$$C + FeO = CO + Fe \quad (12)$$
$$Mn + Fe_3O_4 = MnO + 3FeO \quad (13)$$
$$S + 2FeO = SO_2 + 2Fe \quad (14)$$

These reactions must be taken only as types. In case of each of these five foreign elements it is probable that the oxidation may take place either through ferrous oxide (FeO), or magnetic oxide (Fe_3O_4).

The actual reactions are probably much more complicated than these. In particular, I have simplified them by disregarding the fact that each foreign element is initially in a state of combination, the sulphur as sulphide, the phosphorus as phosphide, etc., or at least in a state of solution.

Of the oxidized products of these reactions the oxides of carbon and sulphur (carbonic oxide, CO, carbonic acid, CO_2, and sulphurous acid, SO_2) are volatile and escape immediately. The oxides of phosphorus, silicon, and manganese (phosphoric acid, P_2O_5, silica, SiO_2, and manganous oxide, MnO) separate mechanically from the iron as oil separates from water, and coalesce with any other oxidized substances present to form the slag, which is an opaque, earthy, lava-like, or vitreous mass, chiefly a silicate of iron oxide, lime, magnesia, and alumina, in very varying proportions, together with phosphoric acid under certain conditions. This slag, when molten, floats upon the molten metallic iron, because it is relatively light, and the two therefore are very readily separated mechanically.

42. The desulphurization by a simple process of oxidation is very far from complete. It is often carried further by means of manganese, the sulphide of which, under favorable conditions, rises to the surface of the

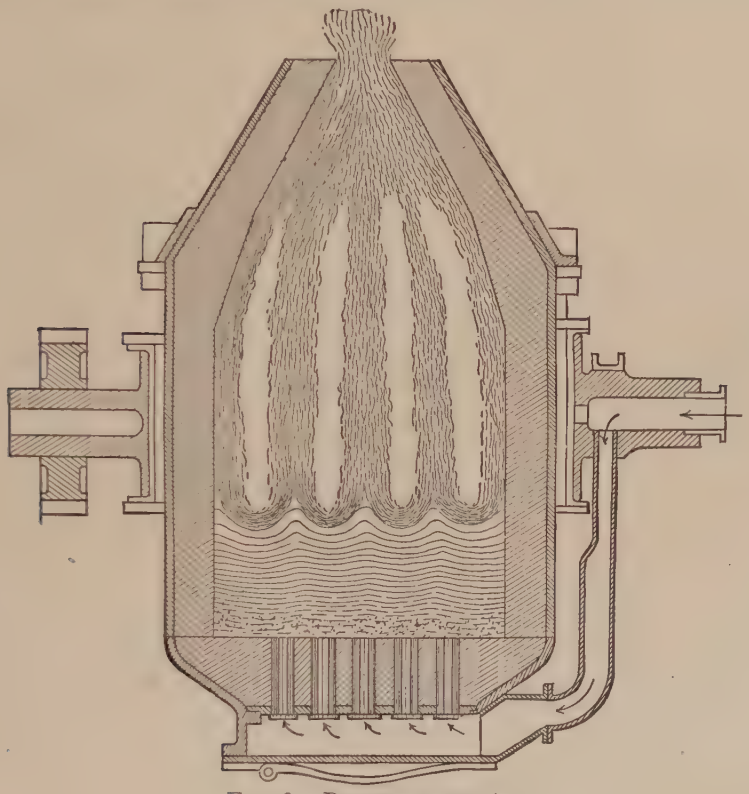

FIG. 6.—Bessemer converter.

molten metal, where its sulphur burns to sulphurous acid, SO_2, and escapes. But if the desulphurization is to be thorough it must be by converting the sulphur into calcium sulphide or its equivalent by reaction (1) already noted,

FIG. 7.—The open-hearth process.

$FeS + CaO + C = CO + Fe + CaS$, and thus removing it into the slag, into which this calcium sulphide passes nearly completely.

The oxidizing reactions are brought about (1) in the puddling process by

stirring iron oxide (in the form of a silicate very rich in that oxide) into the molten cast iron as it lies in a thin boiling layer on the hearth of a reverberatory furnace (Fig. 1, §14); (2) in the Bessemer process by blowing cold atmospheric air through the molten cast iron in a deep clay-lined or dolomite-

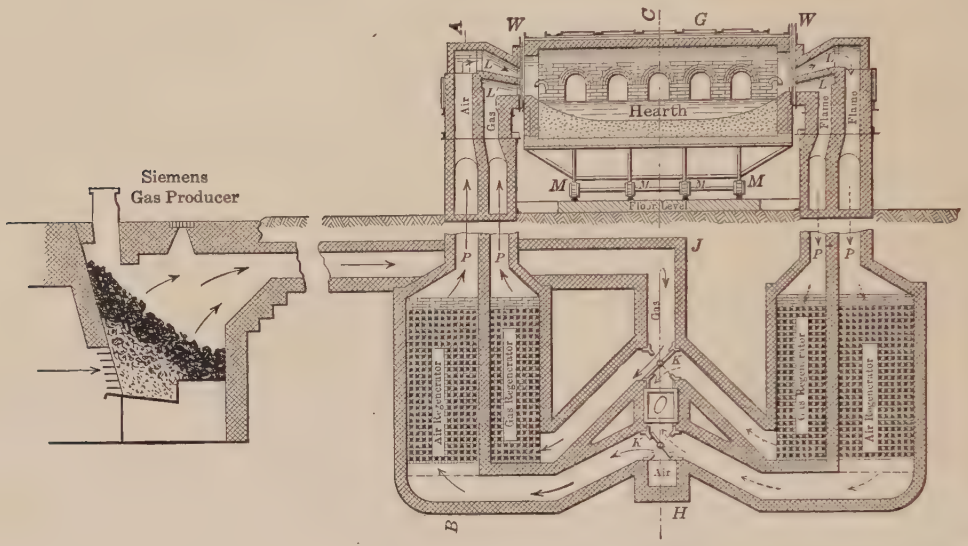

Fig. 8.—Plan through regenerators, flues and reversing valves. Siemens' regenerative gas-furnace for the open-hearth steel process.

Note to Fig. 8.—The metallurgical working part of this cumbersome furnace is that marked "Hearth," and all the rest of the apparatus is for generating the gas, the combustion of which is to heat that hearth, and for preheating that gas and the air needed for burning it. This preheating is done by means of heat traps, filters, or "transfers" called "regenerators," great chambers filled with brickwork piled openly so as to have both an easy path for passing gases, and also enormous surface exposure for absorbing heat from hot gases and giving it up to cold gases.

By means of these regenerators the heat which the products of combustion contain on leaving the hearth is transferred to the gas next to be burnt for heating that hearth, and to the air which is to burn that gas in contact with that hearth, by passing gas, air, and products of combustion, in phases alternately to the right, that is in the direction of the hands of a clock, as shown by the arrows, and in the opposite direction. In the former of these phases the gas and air passing up through the left-hand regenerators preheat themselves by means of the heat which in a prior phase has been extracted from the then escaping products of combustion and impounded there. Thus preheated they meet and burn in contact with the hearth; and the products of their combustion are sucked by the chimney draft out through the right-hand generators, where they deposit their heat. After such a right-hand phase of say 30 minutes, these currents are reversed into a left-hand phase. The right-hand pair of regenerators now preheats the incoming gas and air with the heat just impounded, and the left-hand pair impounds the heat which the present products of combustion contain as they leave the hearth.

Thus each pair of regenerators, during the phase in which it is the admission pair, transfers to the incoming gas and air the heat which during the preceding phase, when it was the exit pair, it took up from the then departing products of combustion.

Ingenious as this system is, it is so cumbrous, and its thermal efficiency is so low, that it never ought to have seemed more than a makeshift. See the Author, "Iron, Steel, and Other Alloys," 2d ed., Sauveur and Whiting, Boston, 1906, p. 350.

lined retort called a converter, the rapidity of the oxidation itself raising the temperature rapidly (Fig. 6); and (3) in the open-hearth and electric processes by exposing the molten cast iron in a thin and very broad layer on

the bottom of a reverberatory furnace to an overlying layer of slag containing iron oxide, and usually enriched in that oxide by throwing into it lumps of iron ore (Fig. 7). In the open-hearth process, in addition to this purification of the cast iron by these oxidizing reactions, its impurities are in most cases also greatly diluted by adding much relatively pure steel scrap.

The open-hearth process bids fair to be simplified and cheapened greatly by the use of powdered coal instead of producer gas. This substitution would enable us to dispense with the extremely expensive Siemens regenerative gas furnace, Fig. 8, which has hitherto been needed for this process.

In order to generate in this furnace the extremely high temperature, about 1,600°C., which the open-hearth process needs, it is necessary to preheat very highly not only the air used for burning the gas, but preferably the gas itself. This pre-heating is done by means of expensive parts called regenerators, Fig. 8, which transfer the heat contained in the products of combustion as they leave the working chamber to the incoming gas and to the air with which that gas is to be burnt.

In most cases this Siemens furnace needs other very expensive apparatus, called gas producers, for generating the gas. The use of powdered coal would not only dispense with these gas producers, but also save most of the cost of the regenerators for transferring the heat from the products of combustion to the incoming gas and air, because the temperature to which the air for burning the powdered coal needs to be raised is so moderate that it can probably be reached economically in the cheap iron-pipe heating stoves. In this case the remaining heat of the products of combustion might be recovered by raising steam.

43. The electric steel-making furnaces[1] are in effect open-hearth furnaces heated by electricity, this being the only thing electrical about the "electric" processes, which are more accurately called "electro-thermal." The metal lies in a molten bath, the shape of which varies very greatly with the method of supplying the heat. It is often supplied by means of an electric arc, as in the Heroult furnace, Fig. 9, in which the current springs as an arc from electrode E to the bath H, passes through the overlying layer of slag, G, and again as an arc passes out through the exit electrode F. In other cases the heat is developed by the electric resistance of the metal itself to a direct current, or more often to an induced current, as in the Kjellin and Roechling-Rodenhauser furnaces.

[1] *Cf. Report of the Canadian Government Commission on Electro-thermic Processes for Smelting Iron Ores*, 1904.

Stansfield, "The Electric Furnace," Toronto, 1907; "Electrothermic Smelting," Ottawa, 1915.

Rodenhauser and Schoenawa, "Electrische Öfen in der Eisenindustrie," Leipsic, Oskar Leiner, 1911.

Idem, "Electric Furnaces in the Iron and Steel Industry," translated by Vom Baur, New York, Wiley, 1913.

Lyon and Keeney, "Electric Furnaces for Making Iron and Steel," U. S. Bureau of Mines, *Bulletin* No. 67, Washington, 1914.

Lyon, Keeney, and Cullen, "The Electric Furnace in Metallurgical Work," U. S. Bureau of Mines, *Bulletin* No. 77, Washington, 1914.

The control over the oxidizing, and more especially over the desulphurizing purification, is far more easy in the electric furnace than in any other, because in it alone are the purifying reactions uncomplicated by the heating conditions.

In the Bessemer process the conditions are violently oxidizing, because only by such violent oxidation can the temperature needed be generated. In the open-hearth process, too, they are naturally strongly oxidizing toward so electro-positive an element as iron, because the atmosphere in the working chamber must needs be very strongly oxidizing in order to permit a fast enough and thorough enough combustion of the fuel there to generate easily the very high temperature which the high melting point of the resultant steel makes necessary. And though, by skillful manipulation, the strength of the oxidizing conditions at the immediate surface of the molten metal, shielded from the atmosphere by the overlying slag, can be mitigated so far as to permit a certain removal of sulphur by the essentially deoxidizing reaction (1) . . . . $FeS + CaO + C = CO + Fe + CaS$, yet it is extremely difficult to reduce the sulphur in this way to below 0.03 per cent., and apparently impossible to reduce it below 0.02 per cent.

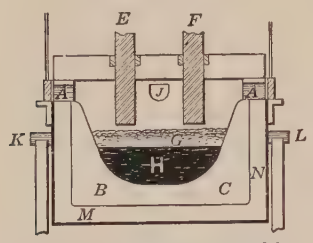

FIG. 9.—Heroult double-arc electric steel purifying furnace.

Because the electric furnace can be given an almost inert atmosphere consisting chiefly of carbonic oxide and nitrogen, it permits a much more accurate control of the strength of the oxidizing or deoxidizing conditions, to meet the needs of the various stages of the purification.

Unfortunately, the great cost of electric heating restricts the use of these electric processes, the natural field for which seems to be to complete the purification begun much more cheaply in the Bessemer converter or the open-hearth furnace. These cruder processes do the major part of the purification far more cheaply, under most conditions, than is intrinsically possible in the electric furnace, which should therefore be used to carry the purification beyond the point at which they are forced to leave it. Beyond this the electric furnace seeks a field of its own in the preparation of certain alloy steels, a field beyond the scene of this bird's eye view.

44. Fining and Recarburizing.—In the Bessemer, open-hearth, and electro-thermal processes the oxidation cannot readily be made thorough enough to remove just the desired quantity of the several impurities initially present, and yet not only to leave just the quantity of carbon desired, but also to leave the iron itself free from oxygen, in general a most hurtful element, but one easily removed in chief part. Hence the purifying or, as it is called, the "fining" phase of each process is usually followed by an extremely brief adjusting or "recarburizing" or "refining" phase, in which the composition is adjusted accurately, chiefly by adding carbon so as to give the desired percentage of that element, and manganese both to give

the needed percentage of manganese and to remove from the iron any oxygen which it may have taken up.

In the crucible process a very small quantity of wrought iron or steel is re-melted in a closed crucible and cast into ingots, or into castings. The desired percentage of carbon is given either (1) by previously carburizing the solid metal by very long heating in contact with charcoal, or (2) by adding charcoal or a pure white cast iron (washed metal) to the charge in the crucible itself. The crucible process is thus in effect a simple re-melting process, with or without simultaneous carburizing, and is thus distinguished sharply from the puddling and Bessemer processes, which are essentially purifying processes. The open-hearth process may be conducted either chiefly as a re-melting process (pig and scrap), or as a purifying process (pig and ore), or it may combine both principles (pig and scrap and ore).

Because the Bessemer process is especially fitted for "roughing" and the open hearth for finishing, these two processes are coming into joint use as a "duplex process," the silicon being removed and the temperature raised in the acid Bessemer converter, and the phosphorus then being removed and the temperature and final composition adjusted accurately in the basic open-hearth furnace, sometimes with treatment in an electric one.

45. Classification of Processes.—We may roughly classify the more important processes as follows:

(1) *The extraction processes,* the blast furnace, and the "direct processes" of making steel or wrought iron direct from the ore; these latter are unimportant today.

(2) *The conversion or purifying processes,* the Bessemer, open-hearth, electro-thermal, and puddling processes. The Bell-Krupp process is one of arrested or incomplete purification. The purification in all these processes consists chiefly in removing by oxidation the excess of carbon, silicon, phosphorus, and manganese introduced in the blast-furnace process, over that desired in the steel or wrought iron.

(3) *The adjusting processes,* adjusting the composition. These include the *carburizing processes,* cementation, case-hardening, and the Harvey and Krupp processes; and the process of making *malleable castings.*

(4) *The Shaping Processes.*—These include the *mechanical processes,* rolling, hammering, wire-drawing, etc., and the *re-melting processes,* those of the iron foundry and the crucible process. The pig and scrap variety of the open-hearth process may, from one point of view, be put here.

Such classifications can rarely be complete or consistent. For instance, while the crucible process as carried out in Great Britain is essentially a re-melting process, as carried out in this country it is at once a re-melting and an adjusting (carburizing) process. Nevertheless these classifications have their use.

CHAPTER 4

CLASSIFICATION AND NOMENCLATURE[1]

Socrates: "But why should we dispute about names when we have realities of such importance to consider?"

Glaucon: "Why, indeed, when any name will do which expresses the thought of the mind with clearness?" Plato: "The Republic."

46. In General.—In this section I give an outline of the classification for those who run, and in §47 and the following sections I consider the subject fully for those who think.

Table 1 gives a general classification of the industrial products.

The most important classes are: Cast iron, wrought iron, and steel.

The classes of second importance are: Malleable castings, special steels, and ferro-alloys.

Cast iron usually contains so much carbon that it is not usefully malleable at any temperature.

Wrought iron contains so little carbon, usually less than 0.12 per cent., that it is very malleable, and that it cannot be hardened greatly by rapid cooling. An additional essential characteristic is that it contains much cinder or slag, because of its "plastic origin," *i.e.*, because it is made by balling up pasty particles of iron in a bath of molten cinder, without later removing it by fusion.

Steel in the *industrial nomenclature*, B of Table 2, used almost universally in the English- and French-speaking countries, includes all malleable kinds of iron except wrought iron and malleable castings. In particular, any malleable kind of iron that contains enough carbon to cause it to harden intensely on rapid cooling is called steel on that account; and any kind that is congenitally malleable, and differs from wrought iron in being of molten origin, is called "steel" on that account. This is true whether it has or has not enough carbon to cause it to harden on rapid cooling. In fact, a very

[1] Greiner, *Rev. universelle des mines*, 1871, No. 1, vol. 33, p. 27; *Journ. Iron and Steel Instit.*, 1914, No. I, vol. 89, p. 29.

Jordan, *Revue de Exposition de Paris*, 1870, vol. 4, p. 280.

Holley, *Trans. Amer. Inst. Mining Eng.*, 1876, vol. 4, p. 138.

Prime, *idem*, 1876, vol. 4, p. 328.

Metcalf, *idem*, 1877, vol. 5, p. 532.

The Author, *Engineering and Mining Journal*, 1875, vol. 20, p. 213; *Trans. Amer. Inst. Mining Eng.*, 1877, vol. 5, p. 515; *idem*, 1908, vol. 52, pp. 928 and 930; "Metallurgy of Steel," 1890, p. 1; *Engineering and Mining Journal*, 1911, vol. 91, pp. 327 and 375; *idem*, 1911, vol. 92, p. 550; *Rassegna mineraria metallurgica e chemica*, March, 1911, p. 135.

The Author and Sauveur, *Report of Committee 24 on Uniform Nomenclature of Iron and Steel to the International Association for Testing Materials, Proc. 4th Congress*, Brussels, 1906; *5th Congress*, Copenhagen, 1909; *6th Congress*, New York, 1912.

TABLE 1.—CLASSIFICATION OF IRON AND STEEL

	Class	Sub-class	Genus	Sub-genus	Species	No.	Current English species names	Varieties
Iron	Normal or carbon series	Malleable	Congenitally malleable[a]	Molten origin	Low carbon[d]	1	Low carbon or mild steel	Bessemer steel, open-hearth steel, crucible (mitis) steel, electric steel.
					Medium carbon	2		
					High carbon	3	High carbon steel	
					Hyper-eutectoid	4		
				Plastic origin[c]	Low carbon[e]	5	Wrought iron	Puddled iron, bloomery iron, etc.
					High carbon	6	Steel	Blister steel, puddled steel (obsolete).
			Post-genitally malleable[b]			7	Malleable castings	White heart, black heart.
		Unmalleable				8	Cast-iron	Gray cast iron, white cast iron, mottled cast iron.
	Special series	Used directly for engineering purposes, etc.				9	Special steels	Nickel steel, manganese steel, chrome steel, high-speed steel, etc.
		Used for introducing alloying elements				10	Ferro-alloys	Spiegeleisen, ferromanganese, ferronickel, ferrosilicon, etc.

[a] Congenitally malleable, as in the case of an ingot or other casting of steel which is malleable as cast, and without further treatment.

[b] Malleable castings, not malleable when cast, but made malleable thereafter.

[c] Aggregated from pasty particles without later fusion.

[d] No exact limits between the low-, medium-, and high-carbon steels have been agreed on. In current American industrial practice the division of the structural steels is about as follows:

Name, for structural purposes	Carbon, per cent.	Tensile strength, lb. per sq. in.
Dead soft open hearth	0.08 to 0.15	50,000 to 60,000
Soft Bessemer	0.08 to 0.10	55,000 to 65,000
Soft open hearth	0.15 to 0.22	
Medium open hearth	0.18 to 0.30	60,000 to 75,000
Hard open hearth	over 0.30	over 70,000

(Tiemann, "Iron and Steel," McGraw-Hill Book Co., Inc., New York, 1910, p. 320.)

[e] With little carbon, usually 0.20 per cent. or less.

large fraction of the steels of commerce, rivet, tube, fencing, sheet, and roofing steel, remain very soft after rapid cooling. They are steels because of their molten origin and not because of any carbon content. The other important classes of steel, rail, axle, tire, spring, and tool steel, are both of molten origin and intensely hardenable by rapid cooling.

In the Philadelphian nomenclature, G of Table 2, used widely in Austria, Germany, and Russia, "steel" is limited officially to alloys of which the tensile strength is greater than 71,000 lb. per square inch (50 kg. per square millimeter), while those less strong, in English "low-carbon steels," are called "ingot iron" (Flusseisen, fer fondu), but ambiguously, because this same name is used generically to include all the molten origin steels. This official limitation is often disregarded by the standard metallurgical writers, and very often in the industry itself.

These two nomenclatures, B and G of Table 2, are identical in all respects, save their names for these congenitally malleable products.

Of the classes of secondary importance, **blister steel** and its derivatives resemble wrought iron in their plastic origin, but differ from it in having enough carbon to make them harden on rapid cooling.

Malleable castings are cast in the state of cast iron, and are then made somewhat malleable by converting their carbon into a relatively harmless state, or removing part of it, or in both ways.

The special steels, such as manganese, nickel, and chrome steel, are steels of molten origin which contain enough of some alloying element, other than carbon, to give them special valuable properties.

The ferro-alloys, such as ferromanganese, ferrochrome, etc., differ from the special steels in containing so much of an alloying element that they are used as vehicles for introducing that element, for instance, into steel or cast iron.

47. The four primary classifications of the industrial iron products are:

(1) By structure, into the eutectiferous and the non-eutectiferous.

(2) By origin, into those of molten and those of plastic origin.

(3) By malleableness, into those which are usefully malleable (steel, wrought iron, and malleable castings), and those which are not (cast iron).

(4) By composition, into the "carbon" series and the "special"[1] series, that is, into the alloys which owe their properties primarily to carbon on one hand and to other elements on the other hand.

Of these classifications the first two represent differences of kind, the third and fourth differences in degree. There is never any question with regard to a given process whether it yields metal of molten or of plastic origin. An alloy either contains eutectic or it does not. The truth of these assertions is not affected by difficulties in determining by what class of process a given alloy of unknown origin has been made, or in deciding microscopically whether

[1] The name "special series" is unfortunate, but less unfortunate than the name "alloy series," which tends to confuse the discussion because all these products are alloys. I use it because I know no better.

TABLE 2.—VARIOUS NOMENCLAT

Ir

Carbon series

Malleable

Congenitally malleable

Molten origin

Plastic origin

Nomenclature										
	A	Pre-bessemer	Non-hardening	Hardening			Non-hardening	Hardening		
			Non-existent	Steel			Wrought iron	Steel		
	B	Present industrial	Steel				Wrought iron			Blister steel, etc.
			Low carbon	Medium carbon	High carbon	Hyper-eutectoid				
	C	Presumptive industrial	Steel							Non-existent
			Low carbon	Medium carbon	High carbon	Hyper-eutectoid				
	D	1912	Steel							Blister metal
			Low carbon	Medium carbon	High carbon					
	E	1914 proposal, initial state	Steel							Blister steel
			Low-carbon steel (ingot iron)	Medium-carbon steel	High-carbon steel	Hyper-eutectoid steel				
				Ingot steel						
	F	1914 proposal, presumptive state	Steel							Non-existent
			Low-carbon steel (ingot iron)	Medium-carbon steel	High-carbon steel	Hyper-eutectoid steel				
				Ingot steel						
	G	Presumptive Philadelphian	Ingot iron				Weld iron			
			Tensile strength				Tensile strength			
			<71,000 lb.	>71,000 lb.			<71,000 lb.	>71,000 lb.		
			Ingot iron	Ingot steel			Wrought iron	Wrought steel		
			Low-carbon ingot iron	Medium-carbon ingot iron	High-carbon ingot iron	Hyper-eutectoid ingot iron				

[1]The division into non-hardening and hardening applies only to nomenclature A, strictly speaking. The

	Special series		Defects	Other considerations affecting fitness for use as the standard
	Used only as vehicles for introducing a special element or elements	Used directly otherwise than as an introducing vehicle		
t able		Non-hardening | Hardening	Based on a useless degree of the hardening power.	
		Non-existent		
		Special steels	1. Two of its "steels," low-carbon steel and blister steel, have nothing in common. 2. Its low-carbon "steel" and some of its special steels are not steel in the pre-bessemer sense.	Its defect (1) would cease automatically on the death of the blister steel industry. It is in consistent use by at least 80 per cent. of those directly interested, who hold it tenaciously.
			Its low-carbon "steel" and some of its special steels are not "steel" in the pre-bessemer sense.	It would succeed the present industrial automatically on the death of the blister steel industry, and thus become automatically the consistent nomenclature of a great majority of those interested.
t n	Ferro-alloys		Substitutes a new name "blister metal" for the present name "blister steel." Its low-carbon "steel" and some of its special steels are not "steel" in the pre-bessemer sense.	The consistency at which it aims would come about automatically on the death of the blister steel industry.
			1. Its low-carbon steel has nothing in common with its blister steel, and 2, is not "steel" according to pre-bessemer usage; 3, its alternative "ingot iron"—"ingot steel" division is inconveniently excentric.	Easiest path to unification, because its terms are now used by both parties, normally by the industrials, and widely for convenience by the Philadelphians.
			1. Its low-carbon steel is not "steel" according to pre-bessemer usage; and 2, its alternative "ingot iron"—"ingot steel" division is inconveniently excentric.	Is a modification of nomenclature E which would come spontaneously on the death of the blister steel industry.
		Special ingot irons	1. Its substantival strength classification is inconvenient; 2, ambiguous in giving "ingot iron" two meanings; 3, not strictly consistent with pre-bessemer usage in calling non-hardening special steels "steel;" 4, its strength division is inconveniently excentric; 5 and 6, its application to the special steels is complex and of doubtful appropriateness.	Used fragmentarily, stumblingly and apparently like a foreign language by most of the small minority which has adopted it. Few if any of them use it consistently, but for several purposes most of them use the industrial nomenclature. Differs from the industrial in such a way that its use would spread slowly.
		Tensile strength		
		<71,000 lb. | >71,000 lb.		
		Special ingot iron | Special ingot steel		

omenclatures, B to G, are not based on the hardening power.

(*Between pages* 38 *and* 39.)

it does or does not contain eutectic. But only an arbitrary line can be drawn between useful and useless malleableness, or between the carbon and the special series, because nearly all of the industrial carbon-iron alloys contain for instance important quantities of manganese, between which and the larger quantities entitling an alloy to be called a "special steel" there is no natural dividing line.

TABLE 3.—FIRST APPROXIMATION TO A STRUCTURAL CLASSIFICATION OF THE IRON ALLOYS

Carbon series	Eutectiferous (approximately cast iron)	Hyper-eutectic cast iron	
		Hypo-eutectic cast iron	
	Non-eutectiferous (approximately steel)	Hyper-eutectoid steels	
		Hypo-eutectoid steels	Pearlitic steel
			Ferritic steel
Special series	With ferrous structure: Eutectiferous	Hyper-eutectic	
		Hypo-eutectic	
	With ferrous structure: Non-eutectiferous	Austenitic	
		Martensitic	
		Pearlitiferous	Hyper-eutectoid
			Hypo-eutectoid: Pearlitic
			Hypo-eutectoid: Ferritic
	With non-ferrous structure		

48. The structural classification, shown in Table 3, subdivides the carbon series first into the eutectiferous and the non-eutectiferous, classes which correspond roughly to the industrial classification into the non-malleable (cast iron) and the malleable (steel, wrought iron, and malleable castings).

Second, the eutectiferous into the hyper- and the hypo-eutectic, and the steels into the hyper- and the hypo-eutectoid.

Third, the hypo-eutectoid steels into the pearlitic and the ferritic, that is, those in which pearlite or ferrite is the dominant member.

The structural division of the special series has not been made out, but as a first approximation we may divide them into those of ferrous structure, that is, those with structure related to that of the carbon series, and those of non-ferrous structure. There is a certain correspondence between this and the industrial division into the special steels, that is, those used directly in the mechanic arts or otherwise, and the ferro-alloys, that is, those used in metallurgy as vehicles for introducing the alloy elements so called.

The ferrous-structure products, or special steels, may be divided into the austenitic, the martensitic, and the pearlitic, all as in Table 3.

49. Merits of a Structural Classification.—A structural classification is fitted for scientific purposes because its divisions are natural, and each of them, save that into the ferritic and the pearlitic steels, is based on a difference of kind.

The correspondence between its primary division into the eutectiferous and the non-eutectiferous cannot readily be made to correspond closely to the

industrial division into cast iron and steel, because the alloys with between 1.70 and 2.2 per cent. of carbon, which belong structurally with the cast irons because they ought in equilibrium to contain a small quantity of eutectic, are yet so malleable that they are naturally classed industrially with the steels, and their makers would resent calling them by so ill-selling a name as "cast iron."

TABLE 4.—COMPOSITION OF WROUGHT IRON AND WELD STEEL CONTAINING BUT LITTLE SLAG, AND OF STEEL CASTINGS RICH IN SLAG

	No.	Carbon	Graphite	Silicon	Manganese	Phosphorus	Sulphur	Slag and iron oxide	Nitrogen
Wrought iron and weld steels (*a*)	1	0.453	0.024	0.105	0.006	0.065	Trace	0.326	.....
	2	0.328	0.022	0.098	0.005	0.073	Trace	0.192	.....
	3	0.512	0.016	0.098	0.029	0.067	Trace	0.308	.....
	4	0.212	0.013	0.080	0.014	0.074	Trace	0.452	.....
	5	0.248	0.014	0.093	0.016	0.084	Trace	0.376	.....
	6	0.229	.....	0.103	0.019	0.089	0.007	0.388	.....
Steel castings (*b*)	7	0.515	.....	0.29	0.70	0.014	0.036	1.89	0.0040
	8	0.58	.....	0.76	0.30	0.020	0.040	0.53	0.0043
	9	0.53	.....	0.54	0.86	0.031	0.056	0.81	0.0028
	10	0.26	.....	0.31	0.67	0.034	0.044	1.65	0.0038

Note to Table 4.—(*a*) Report U. S. Board on Testing Iron, Steel and Other Metals, 1881, I, p. 223. These analyses were made by no less trustworthy a chemist than Mr. A. A. Blair, who informs me that the determinations of slag and iron oxide, though made some 30 years ago, are trustworthy because he reached them by exactly the same methods that he now uses. On the other hand, he doubts whether the small quantities of graphite which he there reported really existed, in view of the advances in our knowledge since then. Private communication, Nov. 3, 1910. Nos. 1 to 6 were intended for making chain cables, and they seem to have been offered as wrought iron, but the board decided properly that they were steel.

(*b*) Private notes.

50. Molten and Plastic Origin.[1]—The iron of the ore is usually brought

[1] The division by origin is much sharper than the division, used till very lately, into "slag-bearing" and "slagless." Indeed though there is very much less slag in Bessemer and like molten origin steels taken as a whole than in wrought iron and blister steel, yet some Bessemer steels seem to contain much more slag, or at least slag plus iron oxide, than some wrought irons; so that if our definitions rested on the words "slag-bearing" and "slagless," then such slag-bearing Bessemer steels would become "wrought iron," which would be contrary to all usage, both good and bad.

Thus Table 4 gives the slag content of certain steel castings, by makers with whose practice I am not acquainted, as between 0.53 and 1.89 per cent. The iron oxide present is included in the slag. On the other hand, the analyses of wrought irons used in making chain cables, given by the U. S. Board on Testing Iron, Steel, and Other Metals (1881), reported on an average only 0.904 per cent. of iron oxide and slag, and among them are those given in Table 4 with between 0.192 and 0.452 per cent. of these substances. Though several of these have so much carbon that they were called by the board "steel," yet they are evidently what the Philadelphian nomenclature calls "weld" steel, presumably puddled steel, because they are so extremely poor in sulphur and manganese that they were evidently made by the puddling process. Stead gives the slag content of the best wrought-iron steamer shafts and wagon axles as from 1 to 2 per cent., and the greatest slag content of molten origin steel as not more than 0.5 per cent. (*Fifth Report Alloys Research Comm., Inst. Mechan. Engin.*, 1899, p. 77).

to the malleable state of either wrought iron or steel by the easy though roundabout path of first turning it into the highly carburized cast iron, and then oxidizing and so removing most of the carbon of this product by the oxygen either of the air or of iron oxide, either native or artificial. But you cannot oxidize the little carbon in this cast iron without incidentally oxidizing some of the incomparably greater quantity of iron itself. And, though the oxides of carbon escape readily from the mass because they are gaseous, the oxides of iron thus formed cannot be separated unless the mass is kept molten till the oxidation is finished, so that they may rise to the surface of the heavy molten iron as cream rises on milk. Because the older forms of apparatus could not set up a temperature high enough to keep the resultant steel or wrought iron molten, these products were always befouled with a considerable quantity of this iron oxide combined with more or less silica in the form of "cinder" or "slag." Indeed, as Figs. F of Plate 14 and C and D of Plate 26 show, they are conglomerates or concretes of metallic iron with this weak brittle slag, which naturally weakens and embrittles the enclosing metal.

51. A Reason for Using Origin as a Basis of Classification.—The reason why we use origin and not the qualities conferred by origin as a basis for classification is that we see no other ready way of indicating those qualities. The cinder of wrought iron, and of blister steel made by carburizing that wrought iron, forms long rods or sheets in rolled and forged objects. These lessen the strength and ductility of the mass, especially transversely; so that one constant difference between wrought iron and the steels which resemble it, except for lacking its cinder, is that the ratio of its longitudinal to its transverse strength and ductility is much greater than the corresponding ratio for steel. In other words, for given longitudinal strength and ductility wrought iron is far weaker and far less ductile transversely than steel. Again, the presence of this cinder facilitates welding. Mechanically, again, the sheets of cinder tend to arrest deep corrosion, but by difference of potential the cinder tends to hasten corrosion. Hence the conflicting claims of the makers of soft steel and of wrought iron as regards rust-proofness. But none of these differences lends itself readily for use as a basis of classification. Hence we retain origin as a basis, with regret.

52. The division into the malleable and the unmalleable, or more accurately into those with and those without a useful degree of malleableness, is evidently a division by degree, and no agreement has been reached as to where the line should be drawn.

It would probably best be drawn by means of a definite standard of malleableness, rather than by a definite carbon-content, first because the carbon content of the structural division most nearly corresponding, that between the eutectiferous and the non-eutectiferous alloys, shifts with the content of silicon and other elements, and second because the degree of malleableness probably varies very greatly with the conditions of solidification, so that a given cast iron may be brought to various degrees of malleable-

ness by various conditions of solidification. When solidified slowly it may be very graphitic and relatively malleable; when solidified more rapidly it may be graphiteless and much less malleable; and with special conditions of solidification the formation of eutectic may be prevented, the cementite which should have formed eutectic be made to unite with the pro-eutectoid cementite, and the malleableness thereby probably increased.

But an approximate idea of the position of the carbon-content which represents the boundary between steel and cast iron may be given by saying that in the absence of complicating conditions, alloys with more than 2.30 per cent. are rarely malleable enough to be classed with the steels, and those with less than 2 per cent. are rarely brittle enough under normal conditions to be classed with the cast irons.

Even the relatively high-carbon cast iron called gun iron has appreciable malleableness, which can be utilized to a limited degree. For instance, the shape of gun-iron pistons of motor cars is often adjusted by bending. Though this slight malleableness is insignificant beside that of even the highest carbon and hence the least malleable steels, it suffices to show that any division based on malleableness must needs be absolutely arbitrary.

We are forced to qualify the definition of cast iron by limiting it to alloys which are not malleable at any temperature, because some steels, notably some tungsten steels, are not usefully malleable when cold.

53. Congenital and Post-genital Malleableness.—Because malleable castings must be distinguished from steel, and because the only easy distinction is that they are not initially malleable but made malleable after they have been cast, whereas steel is malleable as cast, we have to make this distinction between the congenitally malleable, the steels, and the post-genitally malleable, malleable castings. This is a distinction of kind and not merely of degree.

The reason why we use the genesis of malleableness as a basis of classification is that we see no other ready way of indicating the important differences in properties which this difference in genesis causes. The congenital malleableness of the steel is not only usually great but is persistent, whereas the post-genital of malleable castings is readily lost and is confined to the cold and to moderate temperatures. Malleable castings, though often very usefully malleable in the cold, lose much of their malleableness on heating highly, the graphite re-combining to form cementite. Moreover, their malleableness does not increase with the temperature as that of steel does. Finally, they differ strikingly from the weak low-carbon steels with which they compete industrially in being extremely difficult to weld, because at a welding heat they quickly convert themselves into something approaching true cast iron, by the dissolving of their graphite. But because none of these differences lends itself readily as a basis for classification, we retain the genesis of the malleableness for our basis.

54. The Carbon and the Special Series.—To draw the line between these two series, or between the special steels and the ferro-alloys, would be

extremely difficult, and we may question whether the good which it would do would begin to represent the care which it would need (§ 46). A given alloy may be used both as a special steel and as a vehicle, that is, as a ferro-alloy, and one which today is only a ferro-alloy may tomorrow find some direct application as a special steel, as for instance in the case of the very rich nickel alloys.

55. The Existing Nomenclatures.[1]—There are now two nomenclatures of the industrial iron products which, though alike in other respects, differ in the names by which they call the congenitally malleable products, which are indeed by far the most important of all. For purposes of identification I call these the "industrial" and the "Philadelphian" nomenclatures. The former is in almost universal use in the English- and French-speaking countries, and widely used industrially in most if not all other countries. The latter was proposed in Philadelphia in 1876, and is in wide use in Germany, Austria, and Russia. It makes claims to precision and consistency which create in its favor a presumption so strong as to win it many supporters, especially among those who on one hand value these qualities and on the other hand have not understood the gravity of its defects.[2]

56. Origin of the Present Industrial Nomenclature.—From the fourteenth century, and probably from a much earlier time, the genus iron included three important species, steel, wrought iron, and cast iron.

The species steel owed its value to its hardening power, *i.e.*, to the fact that it could be brought at will by thermal treatment either into a soft, malleable state, in which the desired shape, that of an axe, a sword, etc., could be given it, or into a hard strong state, which gave it great value as a material for tools for cutting and fighting. It was made malleable by slow cooling, worked into shape, and then made hard and strong by rapid cooling from a red heat.

But there were many and useful irons which lacked this Jekyll and Hyde property of being shifted at will from hard to malleable and back. They were clearly not of the species steel: they were *iron*. Certain of these owed their usefulness to the fact that they were malleable, indeed much more malleable and more easily welded than steel. Hence they were currently *wrought* into the shape in which they were to be used. Hence in turn they were usually called *wrought* iron. They were too infusible to be melted and cast into a useful shape. The remaining useful irons owed their usefulness to the fact that they were fusible and hence could be cast into a useful shape. Indeed, it was only by casting that they could be given any needed

[1] For a full discussion of the merits of the industrial and the Philadelphian nomenclatures see the Author, "The Unification of the Nomenclature of the Iron Products," *Trans. Amer. Instit. Mining Eng.*, to appear.

[2] I must admit frankly that for many years I was one of these (*The Engineering and Mining Journal*, 1875, vol. 20, p. 283; *idem*, 1911, vol. 91, pp. 327 and 375; *idem*, 1911, vol. 92, p. 550; *Trans. Amer. Inst. Mining Eng.*, 1877, vol. 5, p. 515; *idem*, 1908, vol. 52, pp. 928 and 930; "Metallurgy of Steel," 1890, p. 1; *Rassegna mineraria metallurgica e chemica*, March, 1911, p. 135).

shape, because they were far too brittle to be wrought into shape. Hence their name *cast* iron.

Here, then, we had steel, hard and soft at will; wrought or malleable iron, the most readily wrought but incapable of being either hardened usefully or cast; and cast iron, readily cast but incapable of being wrought. Of these cast iron was very rich in carbon, wrought iron very poor in carbon, and steel intermediate in carbon content. The nomenclature was rational and for a time sufficient.

But it became wholly insufficient when Bessemer's wonderful invention brought into being and at moderate cost a whole new class of products, medium- and low-carbon molten origin steel, the importance of which was far greater than that of the only molten origin product which then existed, Huntsman's high-carbon crucible steel. They were evidently destined to drive the old plastic origin products out of existence save for a few special uses, for instance those for which their very defect of being rich in cinder is a merit.

The new low-carbon steel did not fit into the old nomenclature, for whereas it lacked the hardening power of the existing steels, it also lacked the cinder of the existing wrought iron. But as a whole the new products were so much more important than all the existing kinds of malleable products collectively that it was rather the existing nomenclature that should be fitted to them, than they to the existing nomenclature.

One extremely important consideration in deciding on the nomenclature was that these molten origin steels, old and new jointly, formed a series of complete solidarity as regards structure, properties, manufacture, and use, a series varying continuously in carbon content, with no natural break. Such a series is evidently an important natural unit, and as such needs a collective substantive name. It is this need that prevents our simply inventing and applying a new substantive name, such for instance as "besson" or the "iron of Bessemer," to the low-carbon steels. It would have been inconvenient to set them apart under a distinctive substantive name, for they belong with the steels.

A collective name could be given to this series of molten origin steels by extending the meaning of "steel" so as to include low-carbon steel, for the medium- and high-carbon Bessemer products evidently belonged under "steel" with the old Huntsman crucible steel.

But this was in effect to shift the basis of division between iron and steel from the old hardening power to origin. It was a shifting parallel to the shifting of the chemical essence of "acid" from the old "sourness" to the new essence of "containing hydrogen replaceable by a metal."

This was the course actually followed,[1] resulting in what I here call the industrial nomenclature, B of Table 2.

[1] Greiner proposed this shifting in 1869. Thus in his "Note sur la definition de l' acier" (*Revue universelle des mines*, 1871, I, vol. 33, pp. 27 to 32), he says: "In a note on the use of steel, read before the association in 1869, I proposed to reserve this name 'steel' for all malleable prod-

This shifting of the basis of division between iron and steel was the easiest possible way of giving the molten origin steels as a whole a single collective name. It caused the minimum of inconvenience and of disturbance of the then existing nomenclature, because with the new "origin" basis of division all the important prior products which had been "steel" remained "steel" and all which had been "iron" remained "iron."

The unfortunate side of this shifting was that, if this basis of division were applied rigidly, it would exclude from the class "steel" two relatively unimportant products, blister steel and puddled steel, quite as a strict application of the new basis of "acid" excludes from the class "acid" phosphoric and silicic anhydrides, which have long been called acids.

This exclusion was not put into force and thus an anomaly arose, the old blister and puddled steels lacking the molten origin because of which the new low-carbon steel was called "steel," and the latter lacking the hardening power because of which the former were called "steel," so that there were two classes of steel with nothing in common.

It was probably felt that this anomaly would be short lived, because these steels of secondary importance would probably soon go out of existence, as puddled steel already has. Indeed there is no good reason to expect that

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ucts of the steel industry obtained in a state of fusion, preserving thus the name 'iron' for all malleable products which have not undergone fusion." He then says that Jordan, the eminent Professor of Metallurgy at the Ecole Centrale, Paris, and a member of the Philadelphia Committee which in 1876 proposed the Philadelphian nomenclature, accepts this shifting of the basis of division, in his *Revue de l'Exposition de Paris*, 1870, No. 4, p. 280.

In 1872 Holley said: "Steel is an alloy of iron which is *cast while in a fluid state* into a malleable ingot." *Journ. Franklin Inst.*, 1872, vol. 94.

Though it would profit little to try to trace the progress of the shifting of the meaning of steel from the basis of hardenableness to that of molten origin, yet in order to show how much more favorable to a reversal of this shifting the conditions were at the time of the Philadelphia attempt in 1876 than they would be today, a few facts may here be given. In his famous Cheltenham paper of 1856 Bessemer throughout refers to low-carbon Bessemer steel as "malleable iron," the usual name for wrought iron in Great Britain, and clearly and repeatedly distinguishes Bessemer "iron" from Bessemer "steel," restricting the latter name to legitimate, *i.e.*, higher carbon steel. In his paper before the Institution of Mechanical Engineers in 1861, he continues this discrimination, giving the tensile strength after working of nine specimens of "Bessemer Iron" as between 63,591 and 82,110 lb. per square inch (44.7 and 57.7 kilo. per square millimeter), and that of seven specimens of "Bessemer Steel" as between 136,490 and 162, 974 lb. per square inch (96.0 to 114.6 kilo. per square millimeter). (Jeans, "Steel," Spon, London, 1880, p. 82.)

In discussing the making and merits of wrought iron and steel rails, in the 1869 *Journal of the Iron and Steel Institute*, the eminently practical iron master Mr. Edward Williams says:

"The rails of today are of two distinct kinds—those made from ingots, and those built up; and it must be admitted that, questions of cost and possible supply not considered, ingot rails are best.

"We are in the habit of describing the two kinds as steel rails and iron rails—a description obviously incorrect—because the Bessemer rails now making have a percentage of carbon much lower than that of steel. Besides, it is, so far as I know, impossible to define when iron ends and steel begins. It will then be as well to call the two divisions ingot rails and piled rails." (Vol. 1, p. 156.)

It is clear that even as late as 1869 there had been no general agreement in the iron trade to call low-carbon steel "steel." Greiner's proposal in 1869, quoted above, is the earliest formal attempt of which I know to shift the essence of "steel" from hardenableness to molten origin.

blister steel, which has already sunk to a relatively low order of importance, will survive many decades. With its disappearance the anomaly will cease.

But till then it will certainly remain a defect of the industrial nomenclature, though a defect palliated by the short expectation of life of the blister steel industry, by the unimportance of blister steel,[1] and by the fact that the small class which has to do with it is extremely expert and in no danger of being confused. The great mass of mankind had no need of knowing of the existence of blister steel and its derivatives. Nevertheless a defect must be recognized frankly as a defect.

On the other hand, the nomenclature thus set up has shown itself to be convenient and unambiguous, and in fact fitted to the present and prospective needs of mankind.

Judged by its fruits, its supplying the imperative need of a convenient collective name for the series of molten origin steels, with no change in the names of pre-existing products, and with a resultant nomenclature which in practice is found so convenient and unambiguous that it is used regularly by all classes in all English- and French-speaking countries, this shifting of the basis of division between iron and steel was a good step. Its defect of creating a temporary anomaly is regrettable, but that is to be weighed against the defects of the other or Philadelphian attempt to meet the situation.

57. The Philadelphian Nomenclature, G of Table 2.—Those outside the new Bessemer industry probably doubted whether the old plastic origin steel would thus disappear and remove the anomaly thus injected into our nomenclature, and many of them may have been antagonized by the way in which the name "steel" was extended to cover the new low-carbon Bessemer product, not by means of discussion and agreement among those interested, but as if those who wished this extension had a right to speak *ex cathedra*. Many resisted, and in 1876 a committee of seven eminent metallurgists assembled at the Philadelphia exhibition tried to bring about a return to the pre-Bessemer division by hardening power, and prepared what I call the Philadelphian nomenclature, which divides the congenitally malleable products into the unhardening ones, to be called "iron,"[2] and the hardening ones to be called "steel," and cross-classifies these by origin into "ingot" iron and steel of molten origin and "weld" iron and steel of plastic origin, approximately as set forth under Nomenclature G of Table 2.

[1] Blister steel forms only 0.0024 per cent. of the steel product of the United States (1912), and 0.09 per cent. of that of Sweden (1911). The production in Great Britain is not returned in the statistics of that country, and even the Secretary of the Iron and Steel Institute (of Great Britain) does not know it.

[2] The report which first enunciated the Philadelphian nomenclature called the iron alloys "which will not sensibly harden and temper" "iron" and those "which will from any cause harden and temper" "steel." *Trans. Amer. Inst. Mining Eng.*, 1877, vol. 5, p. 19, Proceedings of the October, 1876, Philadelphia, meeting. This report was made by an international committee of seven, of whom five, Ackerman, Egleston, Gruner, Tunner, and Wedding, were professors at important mining schools; one, Bell, an illustrious manufacturer and writer; and one, Holley, a brilliant engineer and writer. Of these Holley opposed the report strongly, and in writing. (*Trans. Amer. Inst. Mining Eng.*, 1876, vol. 4, p. 138.)

The initial basis of division in this nomenclature, the hardening power, has since been replaced by tensile strength, products with less than 71,000 lb. tensile strength per square inch (50 kg. per square millimeter.) when annealed being "iron" and those with more, "steel." The reasons for this change are first that actual tests prove that much that was intended to be classed as "iron" does harden very greatly, and second that strength is a better measure of the properties which give these products their real value. The division aimed in effect to separate the low- from the medium-carbon steels. The hardening power and the hardness are ill-fitted for making this division, because neither of these properties is prominently useful in the medium-carbon steels, nor is the absence of either property prominently useful in the low-carbon steels. The former fact is shown by the rareness with which the medium-carbon steels are used in the hardened state. It is idle to say that "their degree of the hardening power is useful," because neither that power nor their hardness when hardened is put to important uses. The usefulness of medium-carbon steel is due to its strength and is limited by its sensitiveness to abuse, that is, to its becoming embrittled by cold distortion, as for instance by punching, shearing, and even by slight deformation in handling or working; and the usefulness of low-carbon steel is due to its immunity toward these forms of abuse, and is limited by its weakness. The useful division is into the immune or abusable and the strong. But here are properties which are measured by means of strength, ductility, and shock endurance, and cannot be measured by either the hardening power or the hardness when hardened, except very indirectly and untrustworthily. And if some one property is to be selected as the basis of division, strength is probably as good as any.

The dual meaning of "ingot iron" should be noted. This term has both a collective meaning including all molten origin steels whether low-, medium-, or high-carbon, and also a specific meaning restricted to low-carbon steels with less than 71,000 lb. tensile strength.

58. Extent and Quality of Use of these Two Nomenclatures.—The industrial nomenclature is in practically universal use in the English- and French-speaking countries, Spain, and Denmark; is used in a majority of cases in Holland and Japan; and very widely in Germany and Austria, industrially and unofficially. In the English- and French-speaking countries, at least, it is used as consistently and unhesitatingly as any other part of the vernacular.

To describe the status of the Philadelphian nomenclature is difficult. Though it has been adopted officially by important German organizations,[1] and is nominally used widely in Austro-Hungarian, German, Italian, Russian, and Scandinavian writings, yet its actual use, at least in German writings, is fragmentary and usually wavering.

[1] The German Association for Testing Materials, the German Society of Iron Masters, and the Management of the Prussian Railroads (*Deutscher Verband für die Materialprüfung der Technik*, *Verein Deutscher Eisenhüttenleute*, and the *Preussische Eisenbahnverwaltung*).

Of the standard German writers some use the industrial nomenclature and parts of the Philadelphian apparently interchangeably for certain purposes,[1] others frankly use the industrial regularly for certain purposes, especially for the special steels, for steel castings, and for crucible steel, all of which they call "steel" even if its tensile strength is less than 71,000 lb. per square inch.[2]

Others evidently intend to abstain wholly from using the industrial nomenclature, yet occasionally slip into it, and even their use of the Philadelphian is only fragmentary,[3] for they habitually avoid using the collective name "ingot iron," and instead either enumerate the two species "ingot iron and ingot steel," or else use some term like "material" or "metal" or "alloy." Most of the standard German metallurgical writers avoid applying the Philadelphian nomenclature to steel castings and to crucible steel, and all so far as I have noticed avoid applying it to the special steels, these three classes of products being called "steel" in accordance with the industrial nomenclature even when of low carbon content.

The Royal Institution of Engineers of Holland officially calls all ingots and castings "steel," and all rolled and forged objects "ingot iron" irrespectively of the carbon content.[4]

So in Germany the works are always called "steel works" even if only low-carbon steel is made in them, and not only special steels, crucible steels, and steel castings but also steel ingots are habitually called "steel" industrially even if they are of low-carbon steel.

59. The Defects of the Philadelphian Nomenclature. Its Ambiguity.— That which has chiefly prevented the efforts made during several decades[5] from bringing this nomenclature into general use in the countries in which it has received official sanction, seems to be the ambiguity of its name

[1] Schrödter, *Gemeinfassliche Darstellung des Eisenhüttenwesens. Verein Deutscher Eisenhüttenleute*, Düsseldorf, 1912, p. 4 (official publication of the Association of German Iron Masters).

[2] Prof. Osann of the Royal Mining School of Clausthal regularly calls steel castings "steel," even when they are of low carbon ("Lehrbuch der Eisen- und Stahlgiesserei," Leipsic, 1912, pp. 365, 372).

Mars in his "Spezialstähle" (Stuttgart, 1912, p. 2) asserts that all molten origin steels even if of low carbon content are and ought to be called "steel."

Phillips in Geiger's "Handbuch der Eisen- und Stahlgiesserei" (Berlin, 1911), though recognizing and generally using the Philadelphian nomenclature, at times calls even very low-carbon steel castings "steel."

[3] Professor Heyn of the Royal Prussian Technical High School, Berlin, and Director of the Royal Testing Institution at Gross-Lichterfelde, in the Martens-Heyn "Handbuch der Maierialenkunde" (Berlin, 1912), though evidently making an effort to use the Philadelphian nomenclature, very often if not usually avoids using "ingot iron" in its collective sense, and uses instead either the enumeration "ingot iron and ingot steel," or "material" or "alloy," repeatedly writing of "Bessemer and Thomas material" instead of calling it "ingot iron," and at times using "steel" as the collective name (vol. IIA, pp. 318, 392, and 393). That is to say, he gives the name "steel" to products with less than 71,000 lb. tensile strength, which though "steel" in the industrial nomenclature are "ingot iron" both specifically and collectively in the Philadelphian nomenclature.

[4] L. Bienfait, private communication, 1913.

[5] Heyn, *Internat. Assoc. Testing Materials*, 6th Congress, New York, 1912, *Proceedings of Section A*, p. 157, line 2.

"ingot iron" in having both a collective and a specific meaning. The existence of each of these meanings makes the use of the word in the other meaning ineffective. If I make all kinds of molten origin steels, both low- and high-carbon, as nearly all makers of such steel do, I would not advertise simply "ingot iron," because a contemplating customer needing medium-carbon steel might suppose that I used "ingot iron" in its specific sense and made only low-carbon steel. On the other hand, if I wished to say that car couplers are made of low-carbon steel I could not convey this meaning by saying that they are made of "ingot iron," because I might be misunderstood as using the term collectively and simply asserting that they are made of some kind of molten origin steel. In the former case the existence of the specific sense prevents my showing that I intend to use the term collectively, and in the latter case the existence of the collective sense prevents my showing that I intend to use the term specifically.

Because "ingot iron" is ambiguous, if I wish you to understand that I refer to all molten origin steels collectively I must either enumerate "ingot iron and ingot steel" instead of saying plain "steel" as we do in English, or else depart from the Philadelphian nomenclature and use some other term evidently collective, such as "material" or "metal" or "alloy." If I wish you to understand that I refer only to low-carbon steel, I must add "low-carbon" to "ingot iron," saying "low-carbon ingot iron" instead of "low-carbon steel" as we do in English.

A minor defect of this nomenclature is the cumbrousness which arises from its attempt to make the name of each product express simultaneously its origin, by the adjectives "ingot" and "weld," and its carbon content by the substantives "steel" and "iron," forcing us for instance to say "medium-carbon ingot steel" instead of "medium-carbon steel" as in the industrial nomenclature.

During much of the period before 1876 the plastic origin steels were important, and in speaking of "steels" there may have been some need of indicating whether it was molten or plastic origin steel that was intended, and hence some justification for cumbering our language by forcing us to use a qualifying adjective every time that we spoke of a product, lest for instance "steel," intended to refer to molten origin steel, be misunderstood as referring to or including plastic origin steel. But that need no longer exists. When one says "steel" he practically always refers to molten origin steel, so insignificant are the plastic origin steels and so extremely small the class of people concerned with them. The need of adding "ingot" to "steel" for the purpose of indicating that molten origin steel is intended is little greater than that of adding "cow's" every time that we order "milk" lest we receive asses' milk, or "living" every time we order a horse lest a dead one be brought.

The only important plastic origin product is wrought iron, which is denoted quite as easily and clearly by this its present name as by the name "weld iron."

60. Summary.—In addition to a cumbrousness designed to meet a need which no longer exists, the Philadelphian nomenclature has the grave permanent defect that its name "ingot iron" for the most important group of products, the molten origin steels collectively, is ambiguous in serving also for one of the most important units of that group, low-carbon steel. Because it is thus ambiguous it leaves both this group and that unit without an effective name, as is shown by the practice of careful writers of deliberately abandoning the Philadelphian nomenclature when they wish to be clear.

In view of these defects we interpret the existing use of the Philadelphian nomenclature as such a half-hearted faltering adoption of an inconvenient tool as can be induced in countries in which there is a veneration for authority hard for democrats to understand.

Against these permanent defects of the present form of the Philadelphian nomenclature is to be weighed the temporary defect of the industrial, that its blister steel and its low-carbon steel have nothing in common.

61. The 1912 proposed nomenclature,[1] D of Table 2, sought to avoid the anomaly of the industrial nomenclature, in having in blister and low-carbon steel two steels with nothing in common, by excluding blister steel and its derivatives from the class "steel," and calling them "blister metal." But the expectation of life of the blister steel industry is so short that it is hardly worth while to attempt a change simply for the purpose of removing an anomaly which will disappear with that industry.

62. The 1914 proposed nomenclature,[2] E of Table 2, represents my attempt to meet the existing situation. Its peculiarity is that it recognizes two alternative and equivalent names for low-carbon steel, this English name itself and "ingot iron." If this peculiarity can be forgiven, then this nomenclature has considerable merits, which might fit it for adoption as the standard for reference, with reasonable hope that it might within a measurable time become also the vernacular.

It recognizes the whole of the industrial nomenclature, and enriches it by adding the collective name "ingot steel" to include all molten origin steels except "low-carbon steel." The permissive use of this name can embarrass nobody, and it may be a convenience.

It recognizes the whole of the Philadelphian nomenclature save the unfortunate collective name "ingot iron," which certainly must be amended or ended if this nomenclature is to come into general use, and for this substitutes "steel," which as a collective name is now used very widely by the nominal users of the Philadelphian nomenclature itself. Indeed the industrial fraction of the would-be users of this nomenclature probably use "steel" about as often as "ingot iron" in this collective sense, in order to escape the ambiguity of "ingot iron" and for other motives.

[1] See *Report of Committee* 24, *on Uniform Nomenclature of Iron and Steel*, Proc. Internat. Assoc. Testing Materials, 6th Congress, 1912, First Section, XI, 4, p. 12.

[2] The Author, "The Unification of the Nomenclature of the Iron Products," *Trans. Amer. Instit. Mining Eng.*, to appear.

This proposed nomenclature thus calls for no change from the present usage of either party, save the change from the unusable collective sense of "ingot iron" to "steel." It has the temporary defect of including blister steel and low-carbon steel under the same name "steel." On the death of the blister steel industry it would lose this defect and take on the "presumptive form," F of Table 2.

But no firm opinion concerning this proposal can be offered now, because it has not yet stood the test of discussion.

63. The carbon content of low-carbon steel is not necessarily more than that of wrought iron, though the public may naturally infer that it is, having in mind the old pre-Bessemer division into cast iron very rich in carbon, the pre-Bessemer steels intermediate in carbon content, and wrought iron poor in carbon. In fact the carbon content of much that is called "low-carbon steel" in English and French, in accordance with the industrial nomenclature, is less than that of some wrought irons. Thus so important a product as lap-welded and butt-welded tube steel contains only about 0.06 to 0.07 per cent. of carbon, whereas some material which has at least been represented to be wrought iron has much more carbon. A few cases are collected in Table 4, p. 40, and six other cases are shown in Table 5.

TABLE 5.—WROUGHT IRONS UNUSUALLY RICH IN CARBON

No.	Authority and reference	Description	Chemical composition, per cent.					
			C	Si	Mn	P	S	Slag
1	Percy, *Iron and Steel*, p. 738.	Russian bar-iron...	0.272	.....	.....	.....	.....	.....
2	Percy, *Iron and Steel*, p. 738.	Russian bar-iron...	0.340	.....	.....	.....	.....	.....
3	Percy, *Iron and Steel*, p. 738.	Swedish iron bar...	0.386	.....	.....	.....	.....	.....
4	Exhibit at Paris Exposition, 1889.	French.....	0.35	.....	0.15	0.025	0.04	.....
5	Cubillo, *Journ. Iron and Steel Instit.*, 1901, No. I, p. 453.	Puddled "balls"...	0.24	0.015	0.005	0.015	0.008	.....
6	Riley, *ibid.*, 1892, No. I, p. 93.	Swedish bar iron...	0.171	0.040	0.072	0.006	nil	nil

64. Polyglot of the Chief Classes of Iron and Steel, Table 6.—This is not intended to be complete, but only to include those varieties which are of some importance. It was prepared by Professor Sauveur and the Author with the aid of Committee 24 of the International Association for Testing Materials and others.

65. Of what Material are Malleable Castings Composed?—Both "black-heart" and "white-heart"[1] castings are called "malleable castings," and the material of which they are composed is called in the trade "malleable," using this word as a substantive, and alas more commonly "mallable!" It is also at times called "malleable cast iron," but chiefly in scientific or at least semi-scientific writings, and rarely in trade parlance.

Not only is the well-fixed name "malleable castings" misleading, because these castings are incomparably less malleable than steel castings, but the

[1] See §39.

TABLE 6.—POLYGLOT OF THE CHIEF CLASSES OF IRON AND STEEL IN EIGHT LANGUAGES[1]

Names							
English	French	Italian	Spanish	German	Swedish	Danish	Dutch
Species							
Cast iron	Fonte	Ghisa	Fundición de hierro	Roheisen if not re-melted, Gusseisen if re-melted	Tackjern Gjutjern	Raajern naar detikke er omsmeltet eller Stöbejern	Gietyzer (ruwyzer)
Varieties							
White cast iron or pig iron	Fonte blanche	Ghisa bianca in pani, Ghisa bianca	Lingote de hierro blanco y fundición de hierro blanco	Weisses Roheisen	Hvitt tackjern	Hvidtjern	witgietyzer (witruwyzer)
Gray cast iron or pig iron	Fonte grise	Ghisa grigia	Lingote de hierro gris y fundición de hierro gris	Graues Roheisen	Grått tackjern	Graatjern Graat Raajern	Grauw gietyzer
Mottled cast iron or pig iron	Fonte truitée	Ghisa trotata	Lingote de hierro truchado y fundición de hierro truchado	Halbiertes Roheisen	Hagelstatt or Halfhwitt or halfgrått tackjern	Spragletjern	wit korreliggietyzer
Pig iron (white, gray, mottled, etc.)	Gueuses de fonte ou fonte en gueuse	Ghisa	Lingote de hierro	Gusseisen Roheisen (in Masseln oder Gänsen), Gusseisen (weiss, grau, halbiert)	Tackjern	Pigjern or Raajern	Ruwyzer (gemeleerd)
Hot metal, or direct metal	Fonte de première fusion?	Ghisa da prima fusione liquida	Metal caliente ó metal directo	Remeisen Roheisen, Gusseisen erster Schmelzung	Tackjern direct gjutet från Masugnen	Smeltet Raajern	Ruwyzer cerste gieting
Basic cast iron or pig iron	—	Ghisa basica in pani	Lingote de hierro básico	Thomasroheisen	—	Thomas Raajern	Thomas ruwyzer
Hematite cast iron or pig iron	—	Ghisa ematite in pani	Lingote de hierro hematites	Hämatitroheisen	—	Haematit Raajern	Haematitruwyzer
Malleable pig iron	—	Ghisa per affinazione	Lingote de hierro para fundición maleable	Schmiedbares Gusseisen	—	Hammerbart-Stöbejern	Smeedbaargietyzer
Washed metal	Fonte épurée	Metallo defosforato	Fundición de hierro refinada ó lavada	Entphosphortes Roheisen	Tvåttad	Affosforet Raajern	Ontphospherd gietyzer
Refined cast iron	Fonte mazée	Ghisa affinata	Fundición de hierro refinada ó depurada	Gefeintes Eisen	Raffineradtjern	Raajern til Hardfriskning	Gezuiverd gietyzer
Charcoal hearth cast iron	Fonte mazée	Ghisa al carbone di legna	Fundición refinada al carbón vegetal, en fuego de afinéria	Herdfrischeisen Holzkohlen Herdfrisch-Roheisen	Tackjern för härdfärskning	Raffineret Raajern	Houtskolen gietyzer
Alloy cast iron	Fontés spéciales	Lega di ghise (ghise speciali)	Aleaciones de hierro colado	Sondergusseisen	Special tackjern	Special Raajern	Special gietyzer
Species							
Malleable castings	Fonte malléable	Pezzi fusi (getti) di ghisa malleabile	Piezas de fundición de hierro maleable	Schmiedbares Gusseisen or schmiedbarer Guss	Aduceradt jern	Hammerbart Stöbejern, Hammerbart Stöbegods	Smeedbare gieting

[1] *Proc. Internat. Assoc. Testing Materials*, V Congress, 1909, VIII, 2, table facing p. 18.

Names							
English	French	Italian	Spanish	German	Swedish	Danish	Dutch
Species							
Steel	Acier	Acciaio	Acero	Stahl	Stål	Staal	Staal
Variety A							
Called steel because cast initially into a malleable mass. 1. Soft or low-carbon steel, or ingot iron	Acier doux, acier extra doux, fer fondu	Lingotto di ferro omogeneo (acciaio dolce)	Hierro fundido	Flusseisen[1]	Götmetall	Staal Blödstaal	Smeltyzer Vloeiyzer
2. Half hard and hard, or medium and high-carbon steel, or ingot steel	Acier fondu, acier mi-dur, acier dur	Lingotto d'acciaio, duro, semi-duro	Acero fundido	Flussstahl[1]	—	Haardtstaal	Vloeistaal
Sub-varieties							
Bessemer steel	Acier Bessemer	Acciaio Bessemer	Acero Bessemer	Bessemer-Flusseisen Bessemer-Flussstahl	Bessemerstål	Bessemer staal	Bessemer staal
Open-hearth steel	Acier Martin-Siemens, acier sur sole	Acciaio Martin	Acero de Solera	Flammofen-Flusseisen Flammofen-Flussstahl	Martinstål	Martinstaal	Frischstaal, Siemens-Martinstaal
Crucible steel	Acier au creuset	Acciaio al crogiuolo	Acero de crisoles	Tiegelflusseisen Tiegelflussstahl	Degelstål or Degelgjutstål	Digelstaal	Kroerzenstaal
Cast steel	Acier au creuset	Acciaio fuso	Acero fundido	Gussstahl	Gjutstål	Stöbestaal	Gietstaal
Steel castings	Moulages d'acier	Pezzi di acciaio colato	Piezas de acero colado	Flusswaren	Stalgjutgods	Staalstobegods, or Facongods	Vormgietstaal
Variety B							
Weld steel, or wrought steel called steel because it is capable of hardening greatly by sudden cooling	Fer fort ou fer dur	Acciaio saldato Acciaio fucinato	Acero Soldado	Schweissstahl or Schweisseisen[1]	Wäll stål, rarely used, no equivalent	Svejsestaal	Welyzer Welstaal
Sub-varieties							
Blister steel, also called cemented and converted steel	Acier poule, acier cémenté, acier de cémentation	Acciaio cementato Acciaio al crogiulo Acciaio a carburazione esterna cementato	Acero cementado	Zementstahl	Blasstål, Brännstål, Cementstål	Blaerestaal or Cementstaal	Cementstaal
Shear steel	Acier rafiné une fois corroyé	Acciaio damascato (acciaio saldato)	—	Schweissstahl	—	Svejsestaal	Welstaal
Puddled steel	Acier puddlé	Acciaio pudellato	Acero Pudelado	Puddelstahl	Puddelstål	Puddelstaal	Puddelstaal
Variety C							
Alloy steels	Alliages à base de fer, aciers spéciaux	Lega di acciaii (acciai speciali)	Aleaciones de Acero	Sonderstahl	Special stål	Special staal	Specialostaal
Species							
Wrought iron (or weld iron, or, in Great Britain, malleable iron)	Fer soudé	Ferro fucinato Ferro saldato	Hierro forjado Hierro soldado	Schmiedeisen and Stabeisen	Smidesjern Lancashire jern	Svejsejern	Welyzer
Varieties							
Puddled iron	Fer puddlé	Ferro pudellato	Hierro pudelado	Puddeleisen	Puddeljern	Puddjern	Puddelyzer
Bloomary or knobbled iron	Fer au bois (obtenu au Bas-Foyer)	—	—	Herdfrischeisen	Lancashire Franche-Comté or Walloon jern	Haerd frisket jern (always refers to Lancashire iron from Sweden)	Puddel of frischyzer

[1] According to Wedding cast metal having a tenacity greater than 50 kg. per sq. mm. should be called Flussstahl, while with a smaller tenacity it should be called Flusseisen. Weld metal with a tenacity exceeding 42 kg. per sq. mm. should be called Schweissstahl and with a less tenacity, Schweisseisen.

name "malleable cast iron" for the material of which they are composed conflicts with our classification, which universally recognizes the essence of cast iron as its lack of malleableness at all temperatures. Hence to speak of malleable cast iron is like saying "malleable unmalleable iron," or "black white." I have considered elsewhere[1] the three other courses which suggest themselves. These are:

(1) To legitimize the use of "Malleable" as a substantive for trade purposes, and to provide no scientific name, on the ground that the hodge-podge of which this trade group consists has no single property in common, and hence has not enough interest to science to justify the creation of a scientific group name. According to this plan the substance of which malleable castings are composed is "malleable," a species coördinate with "wrought iron," "steel," and "cast iron."

(2) To call it "malleablized cast iron," and to regard this as not strictly cast iron but a distinct species, quite as the "malleable" of course (1) is.

(3) To invent some new name for it, such as Réaumine, or the iron of Réaumur, who invented the malleablizing process.

66. Definitions of the Chief Kinds of Iron and Steel.—These also were drawn up by Professor Sauveur and the Author for the International Association for Testing Materials, with the aid of its Committee 24, and especially of Mr. H. H. Campbell.[2] I have here made a few verbal changes. The definitions which I have changed in any important respect are indicated with asterisks.

Alloy Cast Irons.—Those which owe their properties chiefly to the presence of an element (or elements) other than carbon. Rare. These products are usually called "ferro-alloys," *q.v.*

Alloy Steel.—See special steel.

Basic Pig Iron.—In America, pig iron containing so little silicon and sulphur that it is suited for easy conversion into steel by the basic open-hearth process. It is restricted to pig iron containing not more than 1 per cent. of silicon.

In Europe, pig iron containing so little silicon and sulphur that it can be converted into steel easily by the basic Bessemer or basic open-hearth process. Here it generally contains 1 per cent. or more of manganese, 1.5 to 3 per cent. of phosphorus, and not more than 1 per cent. of silicon or 0.1 per cent. of sulphur. In Germany, where it is called *Thomasroheisen*, it must contain more than 1.5 per cent. of phosphorus, and it does not usually contain more than 0.5 per cent. of silicon. If its composition is outside of these limits, the pig iron used in the basic Bessemer or basic open-hearth process is called on the Continent not "basic" but generically "phosphoric pig iron," of course provided that it is phosphoric, as it usually is.

Bessemer Pig Iron.—That which contains so little phosphorus and sulphur that it can be used by itself for conversion into steel by the

[1] *Engineering and Mining Journal*, 1911, vol. 91 (Feb. 18), p. 375.

[2] Proceedings of the 6th Congress, New York, 1912.

original or acid Bessemer process. In America this term is restricted to pig iron containing not more than 0.1 per cent. of phosphorus.

In England this term is restricted to pig iron containing not more than 0.06 per cent. of phosphorus or sulphur.

Bessemer Steel.—Steel made by the Bessemer process, whether its carbon content is high, low, or intermediate.

Blister Bar, Cement Bar, Converted Bar.—Bars of blister steel.

Blister Steel.—Steel of plastic origin made by cementing wrought iron with carbonaceous matter. Also, commercially, such steel when heated and worked into merchant sizes. Most writers have used "blister steel" in the former and broader sense. In Sheffield it is used solely in the latter and narrower sense.

Remarks.—The blister steel of commerce is made by cementing very pure wrought iron with charcoal.

Blown Metal.—The red short metal made by purifying pig iron in the Bessemer converter without subsequently removing the oxygen which it absorbs during that purification.

Carbon Steel.—*Definition.* Steel which owes its distinctive properties chiefly to the carbon as distinguished from the other elements which it contains.

Remarks.—Though among the special or alloy steels some are but moderately malleable, among the carbon steels industrial usage confines the name "steel" to products malleable enough to be rolled or forged into merchant shapes.

Cast Iron.—*Definition.* Iron containing so much carbon that it is not usefully malleable at any temperature. In America, besides the foregoing generic meaning, "cast iron" is used also in a specific sense which excludes "pig iron" (see below) and is restricted to cast iron in the form of castings other than pigs, and to re-melted cast iron suitable for such castings.

Remarks.—The cast iron of commerce is reduced from the ore, usually in the iron blast furnace, in direct contact with solid carbon, in the form of coke, charcoal, or anthracite,* and is then tapped from the furnace in a molten state. It always contains an important percentage of carbon, usually from 2.5 to 4.5 per cent., and in most cases an important percentage of silicon.

There are three chief varieties of cast iron.

Gray Cast Iron.—Relatively soft, and characterized by the presence of sheetlets of graphite, often forming an irregular skeleton. This is the variety used chiefly for engineering work.

White Cast Iron.—Extremely hard and brittle; characterized by having all or nearly all its carbon in the combined state, and by its consequent lack of graphite.

Mottled Cast Iron.—Intermediate between gray and white cast iron.

Cast steel in the iron trade means "crucible steel." Obsolescent and undesirable because it might easily be understood to include other steels which have been cast.

Cemented Steel.—The same as blister steel.

Charcoal Hearth Cast Iron.—Cast iron which has had its silicon and usually its phosphorus removed in the charcoal hearth, but still contains so much carbon as to be distinctly cast iron.

Converted Steel.—The same as blister steel.

Crucible Steel.—Steel made by the crucible process, whether its carbon content is high, low or intermediate.

Ferro-alloys.—*Definition.* Iron so rich in some element or elements other than carbon that it is used primarily as a vehicle for introducing that element in the manufacture of iron or steel.

Remarks.—The ferro-alloys are not usually usefully malleable, and they usually contain more of the alloying element than is desirable in a special steel. With variations in industrial conditions the line between special steels and ferro-alloys must needs shift. Indeed, a substance might simultaneously be special steel in the machine shop and a ferro-alloy in the steel mill.

Gray Pig Iron and Gray Cast Iron.—Pig iron and cast iron in the fracture of which the iron itself is nearly or quite concealed by graphite, so that the fracture has the gray color of graphite.

Hematite Pig Iron.—Originally pig iron made from the hematite ores of England, which happen to be so free from phosphorus and sulphur that the pig iron made from them can be used by itself for the acid Bessemer process. By association it has come to mean any pig iron thus relatively free from phosphorus and sulphur. The term is not used in America, and is undesirable.

Hot Metal or Direct Metal.—The molten cast iron from the blast furnace before it has been allowed to solidify. The term is generally applied to molten metal taken direct from the blast furnaces to the steel-making plant.

***Ingot Iron** (French, fer fondu; German, Flusseisen.)—*Generically* all molten origin steel, that is, all iron alloys cast while molten into an initially malleable mass. Identical with the "molten origin steels" of the industrial nomenclature, B of Table 2.

Specifically, such alloys with a tensile strength less than 71,000 lb. per square inch (50 kg. per square millimeter) when annealed.

The nearest English equivalent is "mild," "low-carbon," or "soft steel," names which are used synonymously. The line drawn between medium- and low-carbon steel in countries using the industrial nomenclature does not coincide exactly with the 71,000 lb. tensile strength used in the Philadelphian nomenclature. Thus for structural steels in the United States this line is drawn at about 65,000 lb. tensile strength.[1] But the usage in this respect is not so fixed that it could not be made to coincide with the Philadelphian by means of an agreement.

"Ingot iron" in both senses is used widely in Austria, Germany, Italy, and Russia, as part of the Philadelphian nomenclature, G of Table 2, but not in the English- and French-speaking countries. *It is applied in its

[1] See Tiemann, "Iron and Steel," 1910, p. 320.

specific sense by the American Rolling Mill Company to an extremely pure product, containing only about 0.02 per cent. of carbon, which forms the basis of many of my experiments. In order to be consistent I call this "A.R.M. Steel." See Note A, p. 624.

***Ingot Steel** (French, acier fondu; German, Flussstahl).—Molten origin steel with more than 71,000 lb. tensile strength per square inch (50 kg. per square millimeter.

In English and French this term is used very rarely, if ever, and there is no equivalent for it, the members of this class being called by their specific names "medium-carbon" and "high-carbon steel."

There is no general agreement as to where the line between medium- and high-carbon steel should be drawn, and the position assigned to it varies considerably. A steel of 0.4 per cent. of carbon would be high-carbon steel to the structural engineer, but might be medium-carbon steel to the spring-maker, and low-carbon steel to some makers of cutting tools.

The American practice as regards structural steel is given in a note to Table 1, p. 37.

Iron Castings.—*Definition.* Castings of cast iron.

Remarks.—They are usually made from molten iron direct from the blast furnace, or by re-melting cast iron in a cupola or other furnace, or in crucibles.

Malleable Castings.—*Definition.* Castings which are cast as cast iron and later made malleable without re-melting.

Remarks.—Commercial malleable castings are first cast as brittle white cast iron, and then made more or less malleable either by converting most of their carbon from the state of cementite into that of temper graphite, or by removing most of it by oxidation, or by both means jointly. In both cases the malleablizing is done by close annealing, usually in contact with an oxidizing agent. Thus there are two classes of commercial malleable castings, of which one owes its malleability to a large removal of carbon, and the other to a large removal of carbon from the outer part of the casting, and to the precipitation of much of the remaining carbon in the free or graphitic state. It is not in accordance with good industrial usage to apply the name "steel" to any product of these malleablizing processes, or to any product made by re-melting cast iron in a cupola or like furnace, and such application is generally held to be fraudulent.

Malleable castings differ from steel (1) usually in containing more carbon, and (2) in being cast into a mass which is not initially malleable. They are not strictly of cast iron, but form a distinct species, coördinate with wrought iron, steel, and cast iron. The usual name for the material of which these castings are composed is "malleable," pronounced "mallable," used as a substantive. Those with some respect for their mother tongue, if asked of what material a malleable casting is composed, would generally use a circumlocution (§39).

Malleable Iron.—The same as wrought iron. Used in Great Britain, but

not in the United States, except carelessly as meaning "malleable castings" (vulgar "malleable").

Malleable Pig Iron.—An American trade name for pig iron suitable for converting into malleable castings through the process of melting, treating when molten, casting in a brittle state, and then making malleable without re-melting. The term should be used with care to avoid confusion. This material is also called in trade in America "malleable iron," but this use should be avoided because "malleable iron" has the older meaning of "wrought iron," a meaning which is well established in Great Britain.

Mixer Metal.—Molten cast iron which has been passed into or through a metal mixer.

Mottled Pig Iron and Mottled Cast Iron.—Pig iron and cast iron the fracture of which is mottled, with white parts in which no graphite is seen, and gray parts in which graphite is seen.

Open-hearth Steel.—Steel made by the open-hearth process, whether its carbon content is high, low, or intermediate.

Pig Iron.—Cast iron which has been cast into pigs direct from the blast furnace. This name is also applied loosely to molten cast iron which is about to be cast thus into pigs, or is in a condition in which it could readily be cast into pigs.

Plate Iron.—A name sometimes applied in Great Britain to refined cast iron.

Puddled Iron.—Wrought iron made by the puddling process.

Puddled Steel.—Steel made by the puddling process, and necessarily slag-bearing. (See Weld Steel.) It differs from wrought iron only in being richer in carbon. It differs from most other steels in containing much cinder. Little, if any, of it is now made.

Refined Cast Iron.—Cast iron which has had most of its silicon removed in the refinery furnace, but still contains so much carbon as to be distinctly cast iron.

Semi-steel.—A vague trade name for various products near the border line between steel and cast iron. Among these are low-carbon cast iron made in the air furnace, or in the cupola furnace by the addition of steel scrap to the charge. Also an illegitimate trade name for malleable castings.

Shear Steel.—Steel, usually in the form of bars, made from blister steel by shearing it into short lengths, piling, and welding these by rolling or hammering them at a welding heat. If this process of shearing, piling, etc., is repeated, the product is called "double shear steel;" if not is it called "single shear steel." These names are applied also to the bars and other merchant shapes made by rolling or hammering these products.

Special Steel.—*Definition.* Steel which owes its distinctive properties chiefly to some element or elements other than carbon, or jointly to such other element and carbon.

Remarks.—Some of the special steels necessarily contain an important percentage of carbon, even as much as 1.25 per cent. There is no agreement

as to where the line between the special steels and the carbon steels shall be drawn.

***Steel** (see carbon steel; special steel; ingot steel; weld steel).—As regards steel there are two distinct nomenclatures, B and G of Table 2, the so-called industrial, used nearly universally in English- and French-speaking countries, and to an important extent in the industries of most if not all other countries; and the so-called Philadelphian, or Wedding's, or Teutonic nomenclature, used very generally in German, Italian, and Russian writings, and widely in the industries of those countries.

There are two kinds of steel, those of "molten origin" and those of "plastic origin."

Steel I.—The "molten origin" steels include all alloys cast while molten into a mass initially malleable at least at some one temperature. In the industrial nomenclature all such alloys are called "steel" irrespective of their carbon content and tenacity. In the Philadelphian nomenclature such alloys are called "steel" ("ingot steel;" French, "acier fondu;" German, "Flussstahl") only if their tensile strength exceeds 71,000 lb. per square inch (50 kg. per square millimeter) when annealed. If it does not they are called "ingot iron," which see. This strength corresponds approximately to

a carbon-content of 0.37 per cent. with a manganese-content of 0.35 per cent.
" " " 0.30 " " " " " " 0.60 " "
" " " 0.25 " " " " " " 0.71 " "

With the exception of "blister steel" and its derivatives all the steels which have any present industrial importance, including the "special steels" (see above) fall under this definition.

*Molten origin metal wholly free from carbon is not an industrial product and has no industrial name. It is "steel" specifically, though "iron" generically. (See Note A, p. 624.)

Molten origin steel made by melting in a crucible is called "crucible steel" that made in an electric furnace is called "electric steel." So with Bessemer and open-hearth steel *mutatis mutandis*.

Steel II. Plastic Origin.—*Definition.* Iron which is aggregated from pasty particles without subsequent fusion; is malleable at least in some one range of temperature; and has a tensile strength of at least 71,000 lb. per square inch (50 kg. per square millimeter) when annealed. In the Philadelphian nomenclature this product is called "weld steel," which see.

Remarks.—Blister steel and its derivatives and a few other high-carbon steels which are unimportant today, are the only present steels covered by this definition.

The chief use for blister steel is for re-melting in crucibles to form crucible steel, but some blister steel is rolled or hammered into shear steel, which see, and used for certain kinds of cutlery, etc.

Steel cast (adjective) consisting of solid Bessemer, open-hearth, crucible or other slagless steel, and neither forged nor rolled: applied to steel castings.

For instance, a "steel cast" gun is a gun which is a steel casting, *i.e.*, which has been neither forged nor rolled. To call it a "cast steel" gun would imply that it was made of crucible steel, to which the term "cast steel" is restricted.

Steel Castings.—Unforged and unrolled castings made of Bessemer, open-hearth, crucible, or any other steel. Ingots and pigs are in a sense castings; the term "steel castings" is used in a more restricted sense, excluding ingots and pigs and including only specially shaped castings, such as are generally used without forging or rolling. They may, however, later be forged, *e.g.*, under the drop press, when they cease to be "castings" and become "drop forgings," or if only part is forged then they are partly forgings and partly castings.

Washed Metal.—Cast iron from which most of the silicon and phosphorus have been removed by the rich ferruginous slags of the Bell-Krupp process, or its equivalent, without removing much of the carbon, so that it still contains enough carbon to be classed as cast iron. The name "washed metal" is extended to cover this product even if its carbon is somewhat below the proper limit for cast iron.

Weld Iron.—The same as wrought iron. Obsolescent and needless.

***Weld Steel** (French, acier soudé; German, Schweissstahl).—Iron with a tensile strength of 71,000 lb. per square inch (50 kg. per square millimeter) when annealed, and in addition slag-bearing because made by welding together pasty particles of metal in a bath of slag, as in puddling, and not later freed from that slag by melting. This term is part of the Philadelphian nomenclature and very rarely used in English- and French-speaking countries. Blister steel and its derivatives are the only weld steels made today to an appreciable extent.

White Pig Iron, and White Cast Iron.—Pig iron and cast iron in the fracture of which little or no graphite is visible, so that their fracture is silvery and white.

Wrought Iron.—*Definition.* Malleable iron which is aggregated from pasty particles without subsequent fusion, and contains so little carbon that it does not harden usefully when cooled rapidly.

Remarks.—Commercial wrought iron, though occasionally made directly from the ore, is usually made from cast iron by the puddling process, by such removal of its carbon and silicon as to convert it, at the high temperature used, into pasty particles, and by squeezing these together in a bath of cinder or slag into a coherent mass which retains permanently an important quantity of that slag.

In Great Britain "malleable iron" is often used as synonymous with "wrought iron," but sometimes at the risk of leaving the lay reader in doubt whether it refers to wrought iron or to malleable castings. Hence I advise against this use of "malleable iron."

CHAPTER 5

OUTLINE OF THE CONSTITUTION OF IRON

The Steel-white-cast-iron or Graphiteless Series

67. Summary.—The various kinds of iron and steel, apart from the alloy steels and the ferro-alloys, form a continuous series which is naturally divided into two parts, the graphiteless or steel-white-cast-iron series and the graphitiferous or cast-iron series, which is in effect white cast iron more or less graphitized. These two series meet in white cast iron, which is thus the boundary between them. The present chapter considers the steel-white-cast-iron series, and Chapter 6 will study the cast-iron series. This present study is confined to the constituents themselves. The ways in which they come into existence and change places with each other, that is their genesis and transformation, are studied in Chapters 8 to 11.

The essential difference between the members of the steel-white-cast-iron series, from electrolytic iron and the softest rivet steels to the hardest white cast iron, in their usually slowly cooled state, is in the proportion in which they contain the two microscopic constituents of which they are made up, the soft ductile ferrite or pure iron, and the hard brittle cementite or iron carbide, Fe_3C, (§70). Pure electrolytic iron is composed solely of ferrite, and the hardest white cast iron is composed chiefly of cementite, which indeed might conceivably form the whole of an extreme white cast iron. Between these two extremes there is an unbroken series of intermediate alloys, not all of which are in actual industrial use, composed of these two substances in varying proportions.

Because all the carbon present exists in this form of cementite, the series may be looked at as varying continuously in carbon content from end to end. Indeed this is the usual industrial point of view. Thus we have our low-carbon, medium-carbon, and high-carbon steels, shading off into the relatively low-carbon white cast irons, and thence continuously to the white irons richest in carbon. The ferrite and cementite are in part structurally free, and in part aggregated in three special structural conglomerates, pearlite (§71) which occurs in all these alloys, and the eutectic (§81) and 1.7 per cent. austenoid (§80), which occur in those with more than 1.7 per cent. of carbon. Hence the natural division of the series into the eutectic-free group, which includes most of the steels, and the eutectiferous group which includes the cast irons (§81).

The proportion of ferrite and cementite, and of their special aggregations, is shown quantitatively in Figs. 10 and 11 (§69). The microstructure of the

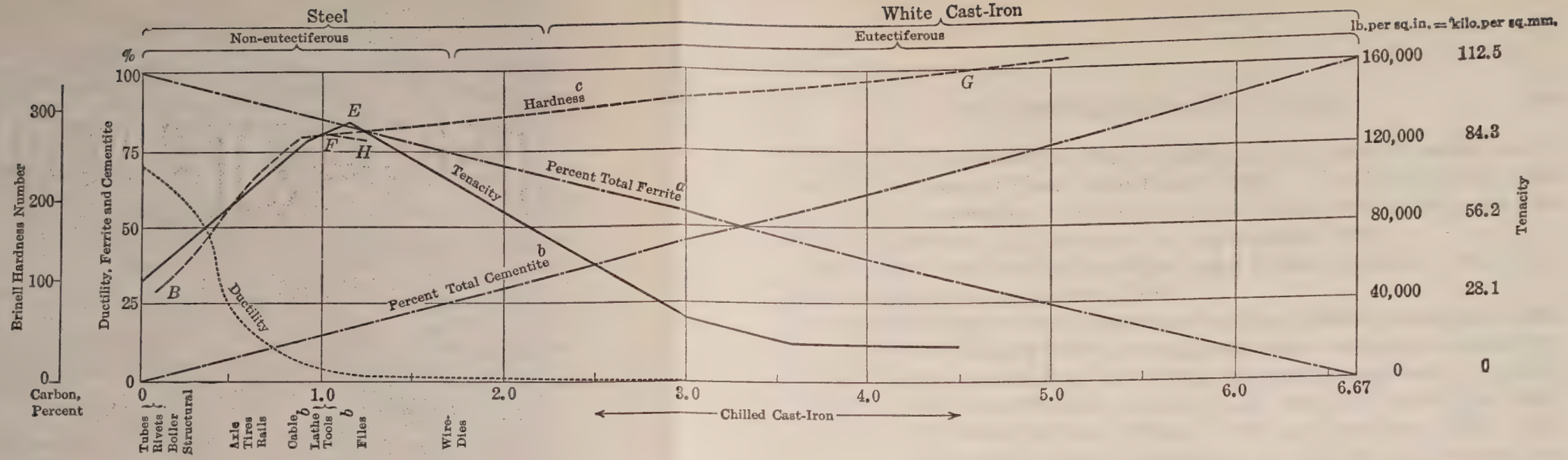
alloys is shown graphically in Fig. 13, (§90) and by means of actual micrographs in the figures of Plate 1 and Plate 3, (§92). The hardening of steel (§97) reflects the presence of one or both of two other constituents, austenite (§95) and martensite (§97) retained by rapid cooling (§97).

These alloys resemble the crystalline rocks in their genesis, constitution, mode of aggregation, and defects (§103).

68. General View of the Constitution of the Carbon Steels and Cast Iron.—Two of the things which most surprise us when we begin to study this wonderful metal, iron, are first that it may contain any and every possible proportion of carbon between zero and say 5 per cent. and even 8 per cent. under special conditions, so that different steels and cast irons differ from each other not by steps as chemical compounds do, but by infinitesimal degrees, not *per saltum* but *per gradum;* and next the extraordinary changes in the properties of these alloys caused by relatively slight changes in either carbon-content or thermal treatment. Both these surprises cease to be surprises when we learn, first, that these irons and steels are not simple substances, but either (1) conglomerates usually of two and sometimes three or even more radically unlike microscopic bodies, exactly as our common granites and like rocks are conglomerates, say of mica, quartz, and felspar; or else (2) solutions, like our glasses; second that certain of these unlike bodies change readily into each other and back, under appropriate manipulations of temperature; and third, that by such temperature manipulations we readily change profoundly the mode of occurrence of these microscopic constituents, the size of the microscopic masses, and the manner in which they are interlaced. The least reflection shows that each of these three kinds of change may be expected to affect greatly the mechanical properties of the mass as a whole, and that the cumulative effect of all three kinds of change brought about simultaneously may well be revolutionary, even without taking into account the great increase of strength, resilience, and hardness which iron, like the other malleable metals, undergoes whenever it is permanently deformed in the cold, whether by wire-drawing, rolling, bending, twisting, or otherwise.

The part played by some of these microscopic constituents is so fundamentally important that we must needs grasp it firmly before we can have an intelligent comprehension of the properties and treatment of steel. Hence a large part of this work will treat of their behavior, structure, and transformations. In particular Chapters 8, 9, and 10 will take a broad survey of their life history, and a later volume will describe them in detail.

69. Industrial Iron and Steel as a Single Carbon-iron Series of Alloys.—Leaving out of sight for the moment the alloy steels and the ferro-alloys (§§46, 55 and 66), let us take at once the broad outlook which embraces all the other important products, low, medium, and high carbon steel, cast iron, and malleable castings, as members of a single series of carbon-iron alloys, varying widely in carbon content as sketched in Fig. 10, from the almost carbonless rivet steel to the most highly carburized cast iron. As



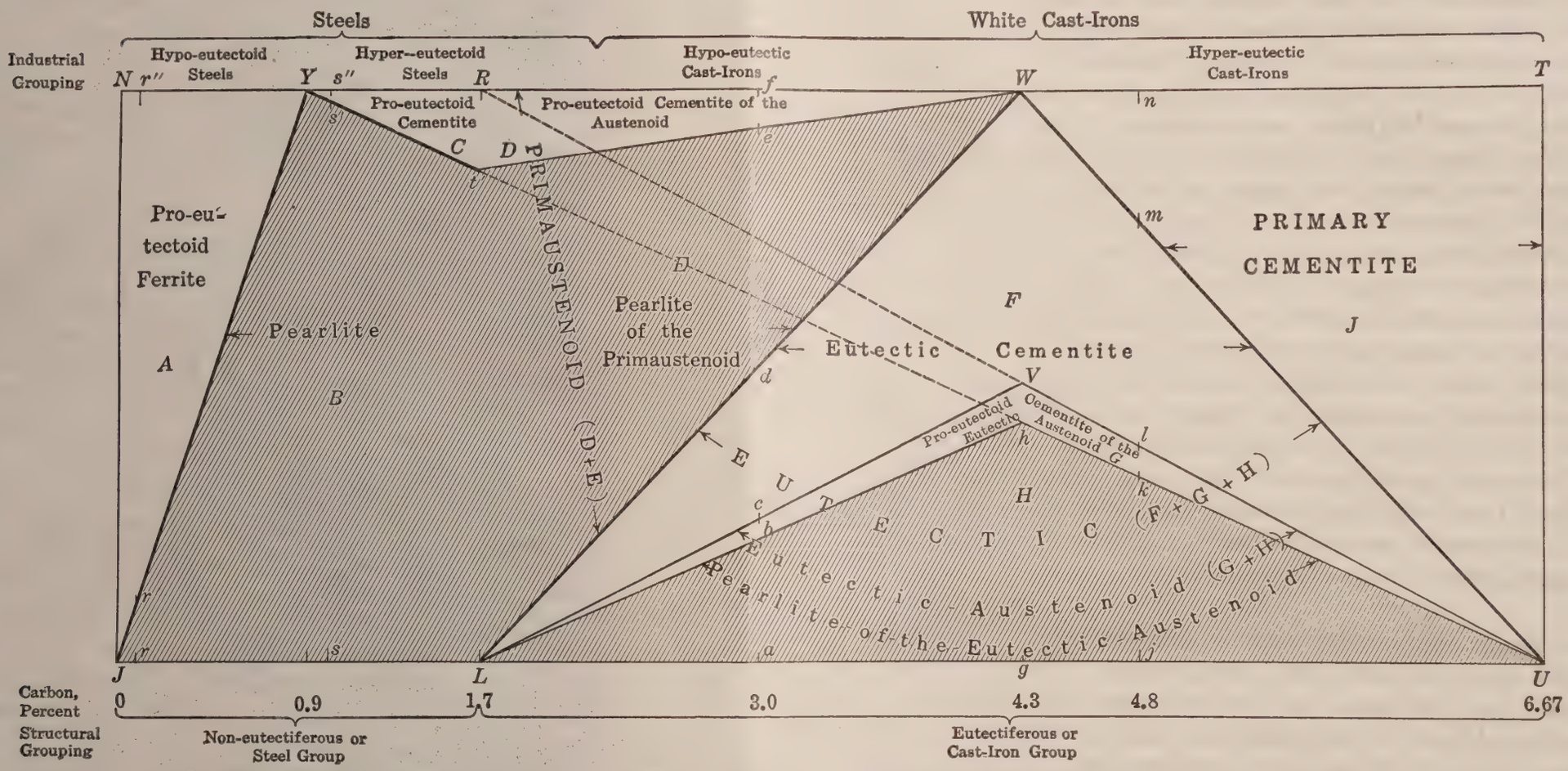


FIG. 10.—Physical properties and assumed microscopic constitution of the pearlitic series, graphiteless steel slowly cooled, and white cast iron.

FIG. 11.—Microstructural constitution of the steel-white-cast-iron series of alloys.

NOTES TO FIG. 10.—*a.* By "Total Ferrite" is meant both that which forms part of the pearlite, and that which is in excess of the pearlite, taken jointly. So total cementite represents the sum of (1) the pro-eutectoid cementite, (2) the pearlitic cementite, (3) the eutectic cementite (*LWUV* in Fig. 11), and (4) the primary cementite. Here too, in hypo-eutectic cast irons the pearlitic cementite includes both that of the pearlite of the primaustenoid, and that of the pearlite of the eutectic austenoid.

b. Lathe tools, files and other cutting tools do not belong to this series, because they, or at least their cutting parts, are used in the rapidly cooled or martensitic state. The purpose of giving their names here is to show how they compare in carbon content with the other objects named.

c. The hardness line from *B* to *H* represents Brinell's determinations. The descent from *F* to *H*, though verified by some others, probably represents slight graphitization. The line from *F* to *G* is hypothetical. The rise here is based on the great hardness of chilled cast iron, and is antecedently extremely probable, because the change of constitution is a progressive substitution of the much harder cementite for the relatively soft pearlite.

NOTE TO FIG. 11.—The use of this diagram is explained at length in §73. In general the constitution of any given alloy is measured on the ordinate passing through its carbon content, by the intercepts between the boundaries of the various areas in the diagram. Thus a hyper-eutectic cast iron of 4.80 per cent. of carbon contains *jm* per cent. of eutectic, and *mn* per cent. of primary cementite; and the *jm* per cent. of eutectic is composed of *lm* per cent. of eutectic cementite, plus *jl* per cent. of eutectic austenoid, of which latter *jk* per cent. consists of pearlite and *kl* per cent. of pro-eutectoid cementite.

This theoretical or equilibrium constitution applies quantitatively only to pure carbon-iron alloys. The quantities of the various constituents are varied by the presence of other elements, such as silicon. In addition to this, in many white cast irons the eutectic is partly or even wholly replaced, jointly by an additional quantity of primaustenoid and by free cementite, due probably to undercooling, as explained in §274.

(*Facing page 63*)

here shown roughly, the ductility is at a maximum and the strength and hardness at a minimum when the carbon content is the least. With increasing carbon content the ductility decreases and the hardness increases, continuously but not at a constant rate, whereas the tensile strength reaches a maximum at a carbon content of about 1 per cent. and thence decreases continuously. Hence the wide variation in the carbon content adapted and adopted for various uses, as indicated below Fig. 10.

Having taken this broad outlook we may note the effects of three important variables other than those of carbon content, A, B, and C.

A. Whereas the great mass of the industrial iron and steel is used in the slowly cooled state, yet for certain objects, such as cutting tools, which need great hardness or strength, the rapidly cooled or martensite state is used.

B. The use of the very rich carbon alloys, cast iron, is not because of, but in spite of their relative weakness and brittleness, and because of their fusibility and hence their ease and cheapness of founding.

C. For most engineering uses of cast iron, this brittleness and the accompanying hardness are mitigated by bringing the carbon, not into the intensely hard brittle combination, cementite, which it forms with part of the iron in white cast irons, but into the soft weak state of graphite, as in gray cast iron and black heart malleable castings.

Thus gray cast iron may be regarded as a sort of offshoot from the high-carbon or white cast-iron end of the steel-white-cast-iron series of alloys, as sketched in Figs. 17 and 18, and explained in Chapter 6, §110.

70. Constitution of the Steel-white-cast-iron or Carbon-iron Series of Alloys: Ferrite and Cementite.—The members of this series, which includes the unhardened steels of which most steel engineering objects consist, and the white cast irons, are composed essentially of two microscopic constituents, ferrite or pure iron, and cementite or the definite iron carbide, Fe_3C, containing 6.67 per cent. of carbon and hence fourteen parts of iron to one of carbon. Thus the carbon is present as cementite chemically combined with fourteen times its weight of iron, and all the rest of the iron is present as ferrite.

Ferrite is soft, ductile, copper-like, and easily welded, whereas cementite is intensely hard, brittle, and glass-like.

The proportion of cementite to ferrite varies continuously throughout this series. The lowest-carbon rivet, tube, and sheet steel, consist essentially of ferrite, for they contain, we may almost say they are contaminated by, only the small quantity of carbon, and hence of cementite, which it is difficult to remove industrially. Their cementite is tolerated rather than sought. Starting from these practically carbonless products there is a continuous increase of carbon content, and hence of cementite content, up to the white cast irons richest in carbon, consisting of about one-third ferrite and two-thirds cementite. For instance, piano or "music" wire, containing as it does 0.75 per cent. of carbon, is a conglomerate of (A), about 11.25 per cent. of cementite, which contains this carbon combined with about 10.50 per cent. of iron, and

(B) of about 89 per cent. of ferrite, which is the whole of the iron not thus combined with carbon in the form of cementite.

This conception of the nature of steel in particular is urged strongly on the reader as essential to a clear understanding of its properties, quite as a recognition of the rôle of the reinforcing metal is to an understanding of the properties of reinforced concrete. Indeed steel and cast iron may well be regarded as a concrete of the ductile ferrite with the brittle cementite, in these widely varying proportions.

From this point of view the fact that the composition of the various classes of iron and steel varies continuously and not by jumps, only reflects the continuously varying proportion of cementite mixed, in one or another of its modes of occurrence, with the ferrite.

Making up as they thus do the whole of the most important series of the iron alloys, and playing a very important part in many of the remaining alloys, ferrite and cementite are so much more important than all the other microscopic constituents jointly that one inclines to call them necessities, and the other constituents hardly more than aids, to such a clear understanding as the engineering practitioner needs. But in this grouping we should include with ferrite and cementite their special modes of aggregation, pearlite and the eutectic about to be described.

This continuous decrease of the proportion of ferrite and increase of that of cementite from end to end of this series is shown graphically in Fig. 10, together with the resultant continuous increase of hardness, continuous decrease of ductility, and increase of tensile strength to a maximum at a carbon content of about 1 per cent., followed by a continuous decrease.

71. Mode of Aggregation of the Ferrite and Cementite in the Steel Group. Pearlite.—The way in which this cementite is associated with the ferrite should be noted. It is interstratified with 6.4 times its own weight of that ferrite, as the peculiar zebra-striped aggregate known as the eutectoid or "pearlite," shown in Figs. A, B and C of Plate 1 and C and D of Plate 2. This pearlite contains about 0.90 per cent. of carbon.[1] Thus a steel of 0.90 per cent. of carbon consists of pearlite alone, Fig. A; one with less carbon consists of pearlite surrounding or surrounded and interpenetrated by the remainder of the ferrite, Fig. B; and one with more carbon consists of pearlite thus surrounded and penetrated by the remainder of the cementite, as shown in Fig. C.

72. Definitions.—These three classes of steel are called "eutectoid," "hypo-eutectoid," and "hyper-eutectoid" respectively. That part of the ferrite and cementite which is contained in the pearlite is called "pearlitic,"

[1] In the case of pure carbon steel the eutectoid carbon content is about 0.90 per cent. But the carbon content of the eutectoid may prove to be affected by the presence of other elements, that carbon content with which pearlite alone, without either additional ferrite or cementite, is formed in the transformation. Steel which thus consists of pearlite alone should be called eutectoid, even if, because of the presence of manganese, or other perturbing element, its carbon content should vary from 0.9 per cent.

and that not contained in it is called "free" or "structurally free" (Sauveur) or "pro-eutectoid" (the Author).

73. The Proportion of Pearlite and of the Pro-eutectoid Element Shown Graphically.—In Fig. 11 the area $JNRL$ represents the steels. Here the ordinate of JYt' measures the percentage by weight of pearlite, and the intercept between JY and Yt' below and NYR above measures the percentage of the pro-eutectoid element.

Thus a hypo-eutectoid steel of 0.10 per cent. of carbon contains rr' or 11 per cent. of pearlite and $r'r''$ or 89 per cent. of pro-eutectoid ferrite. A hyper-eutectoid steel of 1 per cent. of carbon contains ss' or 98 per cent. of pearlite and $s's''$ or 2 per cent. of pro-eutectoid cementite. A hyper-eutectoid steel of 1.70 per cent. of carbon contains Lt' or 86 per cent. of pearlite and $t'R$ or 14 per cent. of pro-eutectoid cementite.

74. The Structure of Pearlite.—On large magnification the zebra-striped or "lamellar" structure of the pearlite itself is seen, as in Figs. A, B, and C, Plate 1, and C, D, and F, Plate 2, with its light and dark curved bands which do not cross each other. In view of the fact that, as usually etched, both ferrite and cementite are white, the presence of these characteristic dark bands in pearlite may surprise us at first. With the usual picric acid etching they represent the unevenness of the etched section, caused by the grooving out of the ferrite, in part by the polishing,[1] but probably in chief part by the etching, leaving the more resistant cementite in relief. Under these conditions the darkness of the dark bands may be referred to shadow effects, and to their being out of focus, covered with an etching patina, and uneven and hence not reflecting their full share of light into the microscope. Hence the width of the black bands need not coincide with the width of the ferrite bands. This helps to explain why the black bands, instead of being some 6½ times as wide as the white ones, that is in the ratio of the ferrite to the cementite by weight, often look but little if at all the wider of the two.

In Figs. C and D of Plate 2 it is the cementite that is black and the ferrite that is white, because this section has been etched with sodium picrate, which blackens the cementite so strongly that the ferrite shows white by contrast in spite of whatever grooving may have occurred in the polishing.

75. Association of Pearlite with Pro-eutectoid Ferrite and Cementite.—In Figs. D and E of Plate 1 the dark masses are the pearlite, which is not resolved into its constituent ferrite and cementite under this magnification.

[1] This grooving out of the ferrite takes place in part in the polishing, but in the usual preparation of microscopic sections most of this darkening effect must be due to the etching, because after polishing and before etching only little darkening is seen. It is true, that by pushing the polishing far beyond that usual in this preparation great relief can be had, and then the contrast between the pearlitic ferrite and cementite is extremely prominent even before etching. Two of the most admirable illustrations of pearlite ever published, showing this banded structure very clearly, are of specimens polished in relief but not etched. (Osmond and Stead, "Microscopic Analysis of Metals," London, 1904, Fig. 17 on p. 35; Benedicks, "Thesis for Doctorate," Upsala, 1904, frontispiece.)

The white masses, including the network, the needles, and the irregular islets, are the pro-eutectoid ferrite in the lower carbon steel, Fig. D, and the pro-eutectoid cementite in the higher carbon steel, Fig. E.

In the great majority of micrographs of steel which are magnified less than 250 diameters the dark masses are of pearlite, which such magnification does not usually resolve, and the white ones the "pro-eutectoid element," whether ferrite or cementite.[1]

The sharp needles in which pro-eutectoid cementite usually occurs suffice in most cases to distinguish it from pro-eutectoid ferrite, which forms thicker blunter masses, so that on inspection we can generally tell at once whether it is a hypo- or a hyper-eutectoid, a low- or a high-carbon steel that we have before us.

76. The Lamellar Structure of Pearlite not always Visible.—Not only is the characteristic lamellar structure, shown in Figs. A, B, and C of Plate 1, usually too fine to be resolved on moderate magnification, but it is often lacking. Thus, when the cooling from the $Ar1$ down, though slow enough to allow the austenite to transform into pearlite, is yet relatively rapid, the individual ferrite and cementite particles are too fine to be resolved by any magnification. The pearlite is then called "sorbite." But beyond this the lamellar structure itself is not so stable but that it can be broken up by the effort of the cementite and ferrite particles to ball up under surface tension, and thus to escape from the lamellar state, the cementite forming sometimes spheroids and sometimes reefs or atolls in a sea of the ferrite, from the embrace of which it has divorced itself. This divorcing effect of surface tension is especially strong when the quantity of either pro-eutectoid element is very large. Thus in very low-carbon steel, with its great excess of pro-eutectoid ferrite, the pearlite divorces with the greatest readiness, and in white cast irons, with their abundant cementite, this divorcing tendency is not only strong but very hard to overcome unless perhaps in the presence of much manganese.

But these variants, sorbite, granular pearlite, divorced pearlite, etc., may for most purposes be regarded as special forms of pearlite, which may be given a generic sense, including them as well as the typical lamellar form.

77. The Carbon Content Inferred from the Quantity of the Pro-eutectoid Element.—The quantity of the pro-eutectoid element seen in a micrograph may give us a rough idea of the carbon content. It is of course zero when the carbon content is 0.90, and it increases in quantity as the carbon content

[1] The relative lightness and darkness of these constituents vary greatly with the details of etching, and in particular with the nature of the etching reagent. The usual reagents of the picric acid class darken the pearlite, leaving the pro-eutectoid ferrite and cementite light. But Rosenhain and Haughton's mixed chloride reagent reverses this contrast for hypo-eutectoid steels, leaving the ferrite dark and the pearlite light (*Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, p. 515). Sodium picrate darkens cementite in whatever form it is present, including the pearlitic cementite, though leaving ferrite light. Hence it reverses the contrast for hyper-eutectoid steels and for pearlite itself, darkening the pro-eutectoid cementite more than the pearlite and the pearlitic cementite more than the pearlitic ferrite. (See Figs. C and D of Plate 2.)

TABLE 7.—THEORETICAL CONSTITUTION OF THE PEARLITIFEROUS SERIES, SLOWLY COOLED CARBON STEEL, AND WHITE CAST IRON

Name		Carbon, per cent.	Proximate composition, per cent.									
			Total pearlite	Pro-eutectoid ferrite	Pro-eutectoid cementite	Primaustenoid	Eutectic	Eutectic austenoid	Eutectic cementite	Primary cementite	Total ferrite	Total cementite
	Low carbon	0.0	0	100	0		0	0	0	0	100.0	0
		0.10	11	89	0		0	0	0	0	98.5	1.5
		0.20	22	78	0		0	0	0	0	97.0	3.0
		0.30	33	67	0		0	0	0	0	95.5	4.5
	Medium carbon	0.40	44	56	0		0	0	0	0	94.0	6.0
		0.50	56	44	0		0	0	0	0	92.5	7.5
		0.60	67	33	0		0	0	0	0	91.0	9.0
		0.70	78	22	0		0	0	0	0	89.5	10.5
		0.80	89	11	0	100	0	0	0	0	88.0	12.0
		0.90	100	0	0		0	0	0	0	86.5	13.5
Steel		1.00	98	0	2		0	0	0	0	85.0	15.0
		1.10	97	0	3		0	0	0	0	83.5	16.5
		1.20	95	0	5		0	0	0	0	82.0	18.0
		1.30	93	0	7		0	0	0	0	80.5	19.5
		1.40	91	0	9		0	0	0	0	79.0	21.0
	High carbon	1.50	90	0	10		0	0	0	0	77.5	22.5
		1.60	88	0	12		0	0	0	0	76.0	24.0
		1.70	86	0	13.9	100	0	0	0	0	74.5	25.5
		1.80	84	0	13.6	96	4	2	2	0	73.0	27.0
		1.90	83	0	13.3	92	8	4	4	0	71.5	28.5
		2.00	81	0	13.1	89	11	5	6	0	70.0	30.0
		2.10	79	0	12.8	85	15	7	8	0	68.5	31.5
		2.20	77	0	12.5	81	19	9	10	0	67.0	33.0
		2.30	76	0	12.2	77	23	11	12	0	65.5	34.5
		2.40	74	0	11.9	73	27	13	14	0	64.0	36.0
		2.50	72	0	11.7	69	31	15	16	0	62.5	37.5
		2.75	68	0	11.0	60	40	19	21	0	58.75	41.25
		3.00	64	0	10.3	50	50	24	26	0	55.0	45.0
White cast iron		3.25	59	0	9.6	40	60	29	31	0	51.25	48.75
		3.50	55	0	8.9	30	70	34	36	0	47.5	52.5
		3.75	51	0	8.2	21	79	38	41	0	43.75	56.25
		4.00	46	0	7.5	12	88	42	46	0	40.0	60.0
		4.30	41	0	6.6	0	100	48	52	0	35.5	64.5
		4.50	38	0	6.1	0	92	44	48	8	32.5	67.5
		4.80	32	0	5.2	0	79	39	41	21	28.0	72.0
		6.67	0	0	0.0	0	0	0	0	100	0.0	100.0

increases or decreases from this quantity. But this indication must be used with very great caution. First, there may seem to be much less of the pro-eutectoid element present than there really is, because it is very often slow in coalescing into masses large enough to be recognized under moderate magnification, especially if relatively little of this element is actually present, as in the case of steels the carbon content of which approaches the eutectoid content of 0.90 per cent. Again the reverse may be true, especially in the

case of very low carbon steels, for their ready divorce decreases the apparent quantity of pearlite and increases the apparent quantity of ferrite, the divorced pearlitic ferrite merging with the pro-eutectoid ferrite.

78. Pearlite in Cast Iron.—Because the white cast irons when cooled slowly contain 1.70 per cent. austenoid (§80) which is equivalent to a steel of 1.70 per cent. of carbon, and because this steel contains 86 per cent. of pearlite as we have just seen, these cast irons contain pearlite.

For like reasons gray cast iron, which is the equivalent of white cast iron with part of its cementite graphitized, usually contains more or less pearlite. This pearlite generally contains much or even all of the cementite which has escaped graphitization.

Typical lamellar pearlite is seen in the gray cast iron of Figs. D and E, Plate 4, and D and F of Plate 13, and in manganiferous white cast iron or spiegeleisen in B of Plate 12. Such good lamellar pearlite as this is rare in spiegeleisen. Divorced pearlite in white cast iron is seen in Figs. D, E, and F of Plate 11.

79. To recapitulate, first the increase of the carbon content from 0 to 0.90 per cent. is accompanied by a continuous replacement of pro-eutectoid ferrite by pearlite, till at 0.90 per cent. of carbon this forms the whole mass as shown from *J* to *Y* in Fig. 11: and then with further increase of carbon content this pearlite is in turn replaced by pro-eutectoid cementite as shown from *Y* to *R* in Fig. 11, till at 1.70 per cent. of carbon the mass consists of 86 per cent. by weight of pearlite and 14 per cent. of pro-eutectoid cementite, in the ratio of 6.1 of pearlite to 1 of pro-eutectoid cementite.

80. Austenoid.—This special complex or conglomerate of pearlite with pro-eutectoid cementite in this 6.4 :1 ratio, or 14 per cent. of pro-eutectoid cementite to 86 per cent. of pearlite, of which unhardened steel of 1.70 per cent. of carbon consists, plays a part of such importance in the structure of the eutectiferous group that we may give it the special name "austenoid of 1.70 per cent. of carbon," or "1.70 per cent. austenoid," recalling the fact that it is the product of the metamorphosis or transformation in cooling of the austenite of 1.70 per cent. carbon which forms in the solidification of all alloys containing more than 1.70 per cent. of carbon. The limiting adjective 1.70 per cent. need be used only when austenoid of his specific carbon content needs to be distinguished from austenoid poorer in carbon.

Because the unhardened steels lower in carbon are in like manner derived from the transformation of austenite of their own carbon content, they are austenoids by analogy, differing only in carbon content from this important 1.70 per cent. carbon austenoid. But little is to be gained on insisting on this analogy, and speaking of these steels as austenoids. The reason for creating the name "1.70 per cent. austenoid," is that this substance is one important constituent of most cast irons; and that we need some ready means of distinguishing it from the other structural entities of cast iron, from the eutectic which is composed in part of this same austenoid, from the primary cementite of hyper-eutectic cast irons, and from graphite. There is no cor-

responding convenience in calling the steels austenoids, for they contain no other constituent from which the austenoid needs to be distinguished, but on the contrary consist solely of austenoid in this sense of the word. Hence I shall not refer to them as austenoids, though I recognize this analogy by bracing all the steels with carbon content of 1.70 per cent. or less together in column 7 of Table 7 as containing 100 per cent. of "austenoid," and by drawing in Fig. 11 no line of demarcation to part the pearlite and the proeutectoid cementite of the steels with less than 1.70 per cent. of carbon from those same constituents of the primaustenoid of the alloys still richer in carbon. Hence both "austenoid," and its two sub-names "primaustenoid" and "eutectic austenoid," refer in this work invariably to the 1.70 per cent. carbon austenoid. This might indeed be called "saturated austenoid," but this term "saturated" has unfortunately become identified with steel of the eutectoid carbon content, 0.90 per cent.

81. Mode of Aggregation of the Ferrite and Cementite in the Eutectiferous or Cast-iron Group. The Eutectic.—With further increase of carbon content, this same special 1.70 per cent. austenoid complex persists in this same 6.1 : 1 ratio, but it forms a progressively smaller part of the whole, being replaced progressively by the entry of a new form of cementite, *eutectic cementite*, the quantity of which present in the alloys of this group is measured by the intercept between *LVU* and *LWU* of Fig. 11.

Moreover, this eutectic cementite forms with part of this austenoid a peculiar aggregate comparable with pearlite, and called the *eutectic* or "*ledeburite*."[1] It contains fifty-two parts by weight of cementite to forty-eight parts of 1.70 per cent. austenoid. Its nature and genesis will be considered fully in §§170 to 175. Its honey-comb structure is shown in Figs. F and G of Plate 1 and M of Plate 3.[2] Here the white comb is the eutectic cementite, and the dark filling the eutectic austenoid, the quantity of which in any given alloy is measured by the ordinate of *LVU* in Fig. 11, while the quantity of eutectic is of course measured by the ordinate of *LWU*, the sum of the eutectic cementite and the eutectic austenoid.

82. Replacement of Eutectic by Cementite.—This quantity of eutectic is that which is theoretically due. But in very many cases the eutectic is replaced in part, or even wholly, by cementite. This replacement I refer to undercooling, as will be explained later. Cases of this replacement are considered in §§179 and 274.

83. Primaustenoid.—The remainder of the austenoid not thus honeycombed as part of the eutectic may be called "1.70 per cent. primaustenoid," because it is derived, with but little change of outline, from the transforma-

[1] What we actually see in the cold cast iron is not indeed the eutectic itself, strictly speaking, but the product of the transformation which the eutectic undergoes in cooling. Nevertheless it retains so closely the structure of eutectic that for brevity it is spoken of as "the eutectic," avoiding coining a new name to distinguish this eutectiform mass from the true eutectic as it exists at the moment of solidification and before undergoing any transformation.

[2] In Fig. M of Plate 3 the great white spine, inclined about 15° to the left from the vertical, in the lower left-hand corner, is of primary cementite, about which the eutectic is oriented.

tion of the primary austenite of 1.70 per cent. of carbon in cooling, as has been explained in §80, and will be explained more fully in §§267, 269, 270, and 272. (See footnote to §96.)

The quantity of primaustenoid, measured by the intercept between LW and RW, decreases progressively from 100 per cent. in the 1.70 per cent. carbon steel, ordinate LR, to 0 in a cast iron with 4.30 per cent. of carbon which consists wholly of eutectic and is hence called eutectic cast iron.

84. Variations in the Carbon Content of the Eutectic.—For simplicity the eutectic has thus far been spoken of as if its carbon content were fixed at 4.3 per cent. In fact its carbon content is lowered by the presence of silicon at the rate indicated by Fig. 12, according to Wüst and Petersen.[1] That is to say, whereas the eutectic carbon content of a pure carbon cast iron free from silicon is 4.3 per cent. or point A, that of a cast iron containing in addition a little more than 1 per cent. of silicon is only 4 per cent., or point E.

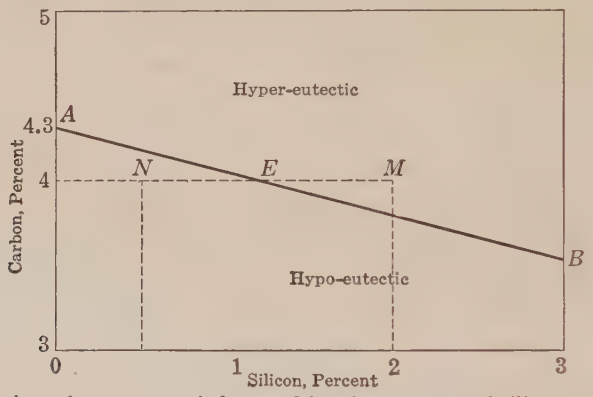


FIG. 12.—The eutectic carbon content is lessened by the pressure of silicon. Sauveur's diagram[2] giving the results of Wüst and Petersen.

With this specific silicon content cast irons are eutectic, hyper-eutectic, or hypo-eutectic according to whether they contain just 4 per cent. of carbon, or more, or less.

TABLE 8.—NATURAL DIVISIONS OF THE STEEL-WHITE-CAST-IRON SERIES WHEN UNHARDENED

Carbon content, per cent.	The cementite in which the carbon occurs is	Name of the alloy	Name of the group
0 to 0.90	Pearlitic	Hypo-eutectoid steel	Non-eutectiferous or steel
0.90 to 1.70	Pearlitic + pro-eutectoid	Hyper-eutectoid steel	
1.70 to 4.30	Pearlitic + pro-eutectoid + eutectic	Eutectiferous steel and hypo-eutectic cast iron	Eutectiferous or white cast iron
4.30 to 6.67(?)	Pearlitic + pro-eutectoid + eutectic + primary	Hyper-eutectic cast iron	

85. Primary Cementite.—With further increase of the carbon content beyond 4.3 per cent. comes still another form of cementite, called primary (§190), the quantity of which is represented by the triangle WTU, Fig.

[1] *Metallurgie*, 1906, vol. 3, p. 811.

[2] Sauveur, *Trans. Amer. Inst. Mining Engineers*, 1913, vol. 45, p. 367, Fig. 6.

11, and in any given white cast iron is measured by the intercept between WU and WT, for instance by mn in the 4.80 per cent. carbon cast iron.

A summary of the carbon condition in these various divisions is given in Table 8.

86. The general parallelism between the eutectic ledeburite and the eutectoid pearlite should be noted. Just as pearlite consists of white masses of pearlitic or eutectoid cementite interstratified with darker ones representing the pearlitic or eutectoid ferrite, so ledeburite consists when cold of white masses of eutectic cementite alternating with darker ones of eutectic austenoid. As we shall see in §170, when first formed it consists of these same cementite masses, alternated with eutectic austenite, which in cooling transforms into this eutectic austenoid.

Again, as the carbon content increases from left to right of Fig. 11, first the pearlite increases from zero to 100 per cent. and then decreases, and next the ledeburite in like manner increases from zero to 100 per cent. and again decreases to zero, so that the pearlite or eutectoid peak Y at 1.70 per cent. of carbon corresponds to the ledeburite or eutectic peak W at 4.30 per cent. of carbon.

Thus pearlite and ledeburite are alike in that each has (A) constant composition, and (B) a constant genetic temperature at which it comes into existence, independently of the carbon content of the alloy in which it occurs, as we shall see in §218; (C) in that neither is in simple atomic proportions; (D) in that the quantity of each increases from zero to 100 per cent. of the alloy, and again decreases, as the carbon content increases from one end of the series toward the other;[1] and (E) in that each is a mechanical mixture or conglomerate of those two substances, ferrite and cementite in the case of pearlite, and austenoid and cementite in the case of ledeburite as we see it in cold cast iron, which substances are the very ones of which one or the other accompanies the pearlite or ledeburite in alloys of lower or higher carbon content. For the ferrite and the cementite which accompany pearlite in hypo- and hyper-eutectoid steels respectively are identical in composition with the ferrite and cementite which jointly constitute pearlite; and the primaustenoid and the primary cementite which accompany ledeburite in hypo- and hyper-eutectic alloys respectively are

[1] The correspondence in this respect is actually closer than that here sketched, because in fact the decrease of the pearlite is continued down to zero at 6.67 per cent. of carbon, as is suggested by the dotted line YU, the ordinate of which measures the total quantity of pearlite present, that of the primaustenoid and that of the eutectic austenoid jointly. Thus the area JYU really matches the area LWU, in that each starts from zero, rises to 100 per cent., and falls again to zero. For the present purpose it is more instructive to divide the pearlite of cast iron up into two parts, that which forms part of the primaustenoid, represented by the triangle $Lt'W$, and that which forms part of the eutectic austenoid, represented by the triangle LhU. In fact these two austenoids, though identical in constitution, yet if regarded as constituents of the structure of the alloy as a whole, are every whit as distinct from each other as the pearlitic and the pro-eutectoid ferrite are. The sum of these two triangles is equal to $Lt'U$.

So, too, the total pro-eutectoid cementite, that of the primaustenoid and that of the eutectic austenoid jointly, is represented by the triangle $YU'R$. Here again it is more instructive to represent these two lots of pro-eutectoid cementite separately, by the areas $Yt'W$ and $LVUh$.

identical in composition with the austenoid and the cementite which jointly make up the eutectic in ledeburite in the cold alloy (§§178 and 190).

The fact that the eutectic, ledeburite, itself contains about five times as much carbon as the eutectoid, pearlite, 4.30 per cent. against 0.90, explains why the eutectic cementite is so much more prominent than the pearlitic cementite. Thus the eutectic contains rather more cementite than austenoid, whereas in pearlite the cementite is in a small minority, only 13.5 per cent. of the whole. In general harmony with this greater quantity of cementite in the eutectic than in pearlite is its enclosing the eutectic austenoid of the former as a comb encloses honey, whereas in pearlite it is simply interstratified with the pearlitic ferrite.

The difference between ledeburite and pearlite, that the formation of the former cannot be prevented by rapid cooling as that of the latter can, is a necessary result of the attendant conditions at birth, as will be explained in §218A.

87. Graphical Illustration of the Proportion of the Various Constituents Present in the Eutectiferous Group.—Just as the proportion of pro-eutectoid cementite present in the austenoid which forms the whole of a steel of 1.70 per cent. of carbon is measured by the intercept $t'R$ on ordinate LR of Fig. 11, so the proportion of this same cementite present, reckoned in percentage of a given white cast iron taken as a whole, is measured by the intercept ef, the triangle LRW representing the decreasing quantity of primaustenoid with increasing carbon content from 1.70 to 4.30 per cent., the triangle $t'RW$ representing the proportion of this primaustenoid which is made up of pro-eutectoid cementite, and the triangle $Lt'W$ the proportion made up of pearlite, or the primaustenoidal pro-eutectic cementite and the primaustenoidal pearlite respectively. In any given white cast iron the percentage of each of these constituents is measured, on the ordinate passing through the carbon content of the cast iron, by the intercept or length of that ordinate between the lower and upper sides of the triangle representing that constituent.

The eutectic austenoid, because like the primaustenoid it is identical in composition with a steel of 1.70 per cent. of carbon, in turn is made up of 14 per cent. of pro-eutectoid cementite and 86 per cent. of pearlite. Thus with increasing carbon content in the eutectiferous group, from 1.70 to 6.67 per cent. of carbon, as the quantity of primaustenoid decreases from 100 per cent. at 1.70 per cent. of carbon to 0 at 4.30 per cent., its percentage of the whole alloy is given by the triangle LRW, as already shown. The quantity of eutectic increases from 0 at carbon 1.70 per cent. to 100 per cent. at carbon 4.30 per cent., and again decreases to 0 at carbon 6.67 per cent., and is represented by the triangle LWU. The quantity of primary cementite increases from 0 at carbon 4.30 per cent. to 100 per cent. at carbon 6.67 per cent., and is represented by the triangle WTU.

Because the constitution of the eutectic itself is fixed, these variations in (1) its quantity imply corresponding variations in the quantity of (2) the

cementite and (3) the austenoid which jointly make up that eutectic, and in the quantity of (4) the pearlite and (5) the pro-eutectoid cementite which jointly make up that eutectic austenoid. Here again the proportion of each of these constituents is measured by the intercept of the area of Fig. 11 which represents it. The area of each of these last five constituents consists of a pair of similar triangles set base to base at the eutectic ordinate, the left-hand triangle of each pair representing hypo-eutectic cast iron and the right-hand one representing hyper-eutectic cast iron.

The sum of the two lots of pro-eutectoid cementite, that in the prim-austenoid plus that in the eutectic austenoid, would in fact be represented by the triangle RUt' drawn in broken lines.

TABLE 9.—AREAS IN FIG. 11 REPRESENTING THE VARIOUS CONSTITUENTS OF THE WHITE CAST IRONS

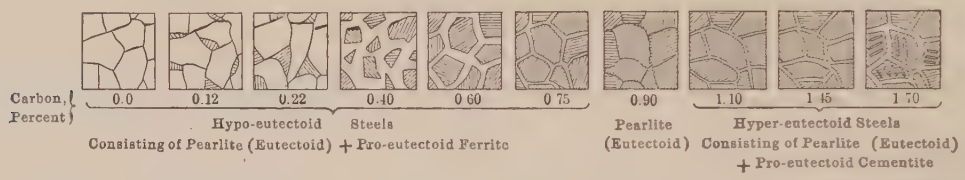
Constituent	Triangles representing it in	
	Hypo-eutectic white cast iron	Hyper-eutectic white cast iron
Primaustenoid	LRW	....
Pearlite of the primaustenoid	Lt'W	....
Pro-eutectoid cementite of the primaustenoid	Rt'W	....
Eutectic	LWg	gWU
Eutectic cementite	LWV	WVU
Eutectic austenoid	LVg	gVU
Eutectic austenoidal pearlite	Lhg	ghU
Eutectic austenoidal pro-eutectoid cementite	LVh	hVU

Examples.—A hypo-eutectic white cast iron of 3 per cent. of carbon, ordinate *af*, Fig. 11, is made up of *ad* or 50 per cent. of eutectic plus *df* per cent. of primaustenoid. Reckoned on the whole alloy, the eutectic contains *ad* or 50 per cent. of eutectic, *cd* or 26 per cent. of eutectic cementite, and *ac* or 24 per cent. of eutectic austenoid, which in turn is made up of *ab* or 20.3 per cent. of eutectic-austenoidal pearlite and *bc* or 3.7 per cent. of eutectic-austenoidal pro-eutectoid cementite. Reckoned on the whole alloy the prim-austenoid contains *de* or 43 per cent. of primaustenoidal pearlite, and *ef* or 7 per cent. of primaustenoidal pro-eutectoid cementite.

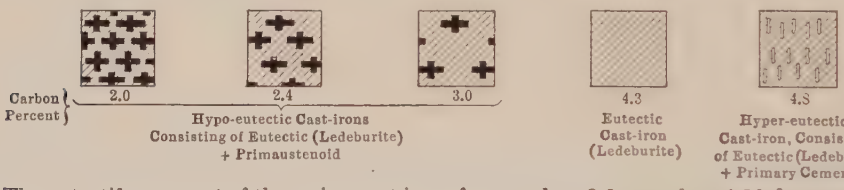
A hyper-eutectic white cast iron of 4.8 per cent. of carbon, ordinate *jn*, is made up of *jm* or 79 per cent. of eutectic and *mn* or 21 per cent. of primary cementite. Of the *jm* or 79 per cent. of eutectic, *lm* or 41 per cent. is eutectic cementite, and *jl* or 38 per cent. is eutectic austenoid, in which in turn there is *jk* or 32.7 per cent. of eutectic-austenoidal pearlite, and *kl* or 5.3 per cent. of eutectic-austenoidal pro-eutectoid cementite.

88. To Recapitulate.—As the steels are conglomerates of the eutectoid, pearlite, with either pro-eutectoid ferrite or cementite according to whether the carbon content is less or greater than 0.90 per cent., so the members of this white cast-iron group are conglomerates of eutectic with either 1.70 per cent. C primaustenoid or primary cementite, according to whether the carbon

content is less or greater than 4.30 per cent. This 1.70 per cent. C primaustenoid is identical with steel of 1.70 per cent. of carbon, and the primary cementite is identical in composition and properties with the pearlitic, pro-eutectoid, and eutectic cementite. To paraphrase the words of §81, first the increase of the carbon content from 1.70 to 4.30 per cent. is accompanied by a continuous replacement of the 1.70 per cent. C *primaustenoid by eutectic*, till at 4.30 per cent. of carbon this forms the whole mass, as shown from *R* to *W* of Fig. 11; and then with further increase of carbon content this *eutectic* is in turn replaced by *primary cementite*, as shown from *W* to *T* in Fig. 11, till at 6.67 per cent. of carbon the mass consists of *primary cementite* alone, or *would theoretically if this sequence could be pushed to its logical conclusion*. It never has been, and perhaps never can be, because of complications beyond our present view.



I, The non-eutectiferous part of the series, steels from carbon 0 to 1.70 per cent., greatly magnified to show the pearlite + (ferrite or cementite) structure.



II, The eutectiferous part of the series, cast irons from carbon 2.0 to carbon 4.80, less magnified, yet enough to show the eutectic + (primary austenite or primary cementite) structure.

FIG. 13.—Diagrammatic sketches of the constitution of the steel-cast-iron series of alloys.

89. Structural Correspondence between the Steel and the White Cast-iron Group.—Note how closely the steel group and the cast-iron group correspond in the sequence of their structural changes which increasing carbon content causes. In particular note that just as the central constituent of steels, pearlite, is a mechanical mixture of ferrite and cementite, which are the constituents which accompany it when the carbon content is either less or greater than 0.90 per cent., so the eutectic is a like mechanical mixture of 1.70 per cent. austenoid and cementite, the constituents which accompany it when the carbon content is either less or greater than 4.30 per cent.

90. Sketch of the Microstructure of the Steel-white-cast-iron Series of Alloys (Fig. 13). The Steels.—Here we trace from left to right in the upper line of sketches the sequence already explained, the progressive replacement of the pro-eutectoid ferrite shown as a white network, by striped kernels of pearlite, as the carbon content increases from 0 toward 0.90 per cent.;

the disappearance of the last of the pro-eutectoid ferrite when the carbon content reaches 0.90 per cent., the whole mass then consisting of pearlite; and the gradual replacement of this pearlite by pro-eutectoid cementite, here forming a new white network, as the carbon content increases from 0.90 to 1.70 per cent., when the 1.70 per cent. C austenoid ratio of 6.1 of pearlite to 1 of pro-eutectoid cementite is reached.

91. The Eutectiferous or Cast-iron Group.—The entry of the eutectic into the series sets up a pattern so strongly marked that henceforth we may conveniently consider the structure as made up of the eutectic masses on one hand, and on the other hand those of primaustenoid or primary cementite as the case may be, leaving out of sight for the moment the fact that the structure of both the primaustenoid and the eutectic austenoid is composite, like that of the 1.70 carbon steel in the right-hand sketch of the upper line. Indeed, both forms of austenoid habitually tend to conceal this composite structure in the presence of the eutectic for reasons which we shall see in §272, and to appear simply as masses of pearlite, irresoluble into its stripes on the scale here used.

The sequence of structural changes with increasing carbon in this group is so closely parallel to that in the steel group that it may be described in exactly the same terms, *mutatis mutandis.*

Here we trace from left to right in the *lower* line of sketches the sequence already described, the progressive replacement of the *primaustenoid,* shown as *black crosses to suggest dendritic or pine-tree growths,* by *honeycombed eutectic,* as the carbon content increases from 1.70 to 4.30 per cent.; the disappearance of the last of the *primaustenoid* when the carbon content reaches 4.30 per cent., the whole mass then consisting of *eutectic;* and the gradual replacement of this *eutectic* by *primary cementite,* here shown as *white blades,* as the carbon content increases from 4.30 to 6.67 when the mass consists of *primary cementite* alone, or would in case this replacement were pushed to completion.

92. The Steel-white-cast-iron Series Shown Micrographically.—Refreshing our memory by looking again at the micrographs already referred to, A to G of Plate 1 and M of Plate 3, we may now examine the whole series shown in Plate 3, starting with carbonless iron in Fig. A and with increasing carbon content passing alphabetically and to the right to very high-carbon cast iron in O. The middle page of this plate shows steels, the right-hand one white cast irons.

93. The Steels.—Note how, as we pass forward alphabetically, the white ferrite, which forms the whole of the carbonless iron, A, is replaced progressively by the dark pearlite, till in the 0.92 per cent. carbon steel, F, the whole mass is of pearlite; and how with further increase of carbon content the dark pearlite is in turn replaced progressively by the white pro-eutectoid cementite. This latter replacement is at most very incomplete, because even with 1.70 per cent. of carbon there is still 86 per cent. of pearlite.

94. The White Cast Irons, Figs. J to O, Plate 3.—On the right-hand page of Plate 3, Figs. J to O, we can trace the sequence of the structural replacements with increasing carbon content already described in §91. This sequence parallels that of the steels so closely that a paraphrased description of the latter suffices for the former, thus: Note how, as we pass forward alphabetically, the *black austenoid* is replaced progressively by the *white eutectic* till, in the *approximately eutectic area of Fig. M*, the whole mass is of *eutectic;* and how with further increase of carbon content the *eutectic* is in turn replaced progressively by the white *primary cementite*, as in Figs. N and O.[1]

The eutectic is not resolved into its component austenoid and cementite by the small magnification of Figs. J and K, and here shows white. With the larger magnification of Figs. L to O, its honeycombed structure is evident.

In Figs. J and K the dendritic structure of the abundant primaustenoid is very prominent. In Fig. L this dendritic structure is still present though it would hardly be recognized except by a practised eye. Here the large black areas represent the small quantity of primaustenoid present in this nearly eutectic field, and the white and the narrow black bands represent the eutectic cementite and the eutectic austenoid respectively.

The ultra high-carbon steel, Fig. I 2 of Plate 3, is intermediate in carbon content between the highest carbon steel of the middle page, Fig. I, and the lowest carbon cast iron of the right-hand page, Fig. J. Structurally it belongs with the cast irons, because in addition to the pro-eutectoid cementite of the hyper-eutectoid steels it contains also eutectic cementite, that is, cementite which ought normally to be interstratified with austenoid in the typical eutectic or ledeburite grouping. But in fact it has escaped from this intertwining, as indicated in §82. Its identity has been concealed by the pro-eutectoid cementite which has united completely with it, to form the large white masses so conspicuous in this micrograph.

But in spite of its large cementite content this alloy is so malleable as to be classed industrially with the steels.

95. Three other constituents here need notice: austenite, martensite, and graphite.

Austenite, often called "gamma iron" especially in France and "mixed crystals" especially in Germany, is the non-magnetic high-temperature state into which the pearlite, and its accompanying pro-eutectoid ferrite or cementite in hypo- and hyper-eutectoid steel respectively, pass spontaneously and merge themselves on rising through the "transformation range," called often the critical range of temperature. This transformation range will be considered at length in Chapters 9 and 10. Suffice it here to say that, in carbon steels, its lower boundary is at about 700°, while its upper boundary varies with the carbon content, from about 700° to 1,135°, so that while the

[1] This figure O is of interest because it is one of those published by Tiemann in 1901, in which so far as I know primary cementite was first clearly identified. See Tiemann, *The Metallographist*, 1901, vol. 4, p. 322, Fig. A2, and Benedicks, *Metallurgie*, 1906, vol. 3, pp. 426, 431, and 468, and Fig. 10, following p. 432.

range may be extremely narrow as in steel of 0.90 per cent. of carbon, it may be more than 400° wide, as in steel of 1.70 per cent. of carbon. In this transformation the various constituents merge as austenite.

Just as the unhardened steels and white cast irons when cold are mechanical mixtures of ferrite plus cementite partly interstratified as the special aggregates pearlite and eutectic, so the steels when above the transformation range are austenite, and the cast irons on solidifying are a mixture of austenite plus cementite, in part interstratified as the special aggregate, the eutectic.

96. Relation of Austenite to Ferrite.—Thus austenite plays at temperatures above the transformation a part like that of ferrite at all temperatures below that range. It might indeed be likened to a high-temperature nonmagnetic ferrite, differing from true ferrite in containing carbon in solution in proportions varying from 0 to 1.70 per cent. The carbon probably exists in that solution in the state of cementite. Thus ferrite and austenite are alike in representing the whole of the iron present except that which is combined with the cementite, if we add that in the case of austenite this exception obviously does not include that part of the cementite which itself is dissolved in the austenite, but includes all the undissolved cementite. This undissolved cementite comprises in hyper-eutectoid steels the pro-eutectoid cementite, and in cast irons any pro-eutectoid and eutectic cementite, and any primary cementite if there is any. Primary austenite plays at high temperatures a part like that of the pro-eutectoid ferrite of cold steel, and eutectic austenite a part like that of pearlitic ferrite.

Thus austenite is essentially solid iron with carbon dissolved in it probably as cementite in indefinite proportions, up to the saturation point at 1,135°, 1.70 per cent. That is to say, it bears the same relation to cementite that a brine bears to solid salt, save only that the austenite is solid whereas a brine is liquid.

On again cooling slowly through the transformation range the austenite changes back into its initial complex of pearlite with either pro-eutectoid ferrite or pro-eutectoid cementite, according to whether the austenite contains less or more than the eutectoid carbon content, 0.90 per cent. In the case of cast irons I have called this complex of pearlite plus pro-eutectoid element "austenoid,"[1] suggesting its derivation from austenite and its retention of the shape of that austenite (§80).

Steel in the austenite state is said to have the hardening power, as explained in §§97 and 98.

The microstructure of the familiar manganiferous austenite, Hadfield's

[1] In the cooling of cast irons which when hot consist of primary austenite plus eutectic, the 1.70 per cent. C primaustenoid and eutectic into which they transform retain very closely the outline and pattern of the original masses of austenite and eutectic. Hence comes the current practice of speaking of the photomicrographs of such cast irons as showing primary austenite plus eutectic, and as showing eutectic austenite as the black filling of the eutectic honeycomb, in spite of the fact that the substance actually photographed is not austenite but that product of transformation which, to prevent the possibility of misunderstanding, I have called "austenoid." The license is permissible, as it is to say that a photograph of a true bust of Napoleon shows Napoleon's features.

manganese steel, is shown in Fig. B, Plate 39, with its very marked bands caused by permanent deformation in the cold.

97. Martensite, of which hardened steel consists, represents a state intermediate between that of the high-temperature austenite and that of the low-temperature pearlite plus ferrite or cementite, a state in which most carbon steel is spontaneously caught in transit when it is cooled rapidly, as in hardening by quenching in water. That is to say, under the usual obstructing conditions, the transformation from austenite to pearlite plus ferrite or cementite, which occurs in slow cooling, though itself rapid, yet is not extremely rapid, and moreover it cannot take place in the cold except to an extremely slight extent.

98. The Hardening of Steel.—Hence, a very rapid cooling of steel from the austenitic state hurries it through the range of temperature in which this transformation not only tends to occur but can occur, to the cold in which it cannot occur, and so the transformation is prevented from completing itself. Hence the catching of the metal in this intermediate hard brittle state of martensite and the hardening of the steel thus implied. This hardening should be considered essentially as a result of the martensitization, brought about in this individual case by rapid cooling, but also brought about under other conditions by other means and without rapid cooling, and then also causing hardness (§§237 *et seq.*).

The hardness of martensite and unfortunately also its brittleness increase with its carbon content, which is one chief reason why the carbon content of tool steel is directly proportioned in practice to the hardness needed in the tool and to the degree of brittleness which can be endured, files containing as much as 1.50 per cent. of carbon. Lathe tools, which cannot endure such brittleness as is tolerable in a file, usually contain about 1 per cent. Not less than 90 per cent. of what is classed as "tool steel" contains between 0.75 and 1.35 per cent. of carbon.[1]

99. The Structure of Martensite.—The characteristics of pure austenite thus caught can best be explained later, but we may note here the appearance of the martensite, and of the mixture of martensite and austenite, in rapidly cooled steel. The zigzag needles and lance heads of martensite are seen in Fig. A, Plate 2, in a mass of austenite. In this case they occur in a relatively smooth ground mass of austenite, grouping themselves apparently about its cleavages. They are sometimes lighter and sometimes darker than the enclosing austenite. Similar zigzag martensite masses are seen in the very high carbon steel of Figs. A and B of Plate 10, and in the dark areas of the white cast iron of Fig. A of Plate 11, and less clearly in those of Fig. B of Plate 11. As will be explained in §272, these areas in this cast iron are equivalent to a very high carbon steel of about 1.70 per cent. of carbon.

A well-marked martensitic structure is readily had by quenching very high-carbon steel very rapidly from a very high temperature. But this

[1] Mathews and Stagg, "Factors in Hardening Tool Steel" *Trans. Amer. Soc. Mechan. Engin.*, 1914, p. 3 of preprint.

structure becomes less strongly marked as the carbon content, the quenching temperature, and the rapidity of quenching becomes less. Note in Plate 8 that even with the high-carbon steel of Figs. F and G, the martensitic structure, though tolerably well marked in Fig. F after quenching from a higher temperature, is hardly seen in G after quenching from a lower temperature. Here as in many cases the martensite looks much like poorly resolved pearlite. The well-marked martensite of Fig. A of Plate 11, after quenching this cast iron from 1,050°, is replaced by a much less clearly marked martensite in Fig. B after quenching from 765°, and is lacking altogether in C after quenching from 728°.

The question whether martensite is a true constituent, comparable with those others, or only a structure, may be deferred till later. But whether structure or constituent, it is of the first importance.

100. The Hardening Power and Thermal Treatment.—The facts that the ductile conglomerate of which unhardened steel consists, a great mass of soft ferrite with a little hard cementite scattered through it, is thus changed spontaneously into hard brittle martensite, which is far harder than the average of that conglomerate, and that a well-controlled re-heating can again bring the metal back to any desired transition stage between this martensite and the initial state of the ferrite plus cementite conglomerate, explain in large part why the properties of steel are so profoundly affected by thermal treatment. As we have seen, further explanation is given by the variations which such treatment causes in the structure of the ferrite and the cementite, which may at will be massed into coarse masses with extended weak cleavage planes, or into particles of ultra-microscopic fineness. Still further explanation is offered tentatively by the apparently amorphous state into which part of the metal can be brought by cold deformation or overstrain, with great elevation of the elastic limit, as in wire-drawing, and the return thence to the normal or crystalline state by even brief re-heating to about 600°C.

101. Graphite is an essential constituent of all gray cast iron, indeed the source of its grayness. It is pure carbon, soft and weak, habitually forming thin flakes scattered as a discontinuous skeleton through the cast iron which, as a practically foreign body, it weakens and embrittles. This may be understood readily from its appearance in Fig. F, Plate 6, where it forms long, thin, sinuous black lines, most of them from $\frac{1}{32}$ to $\frac{1}{100}$ in. thick, sometimes nearly 2 in. long, as here seen, and often bordered on one or both sides by white curved streaks of ferrite, parallel to the graphite and about $\frac{1}{16}$ in. wide. See also Fig. E of Plate 4. Such graphite flakes are one cause of the inferiority of cast iron to steel, weakening and embrittling the alloy very greatly. Their influence is that of a sponginess, because the graphite itself is as a graphite link would be in a steel chain. Fortunately this damage may be lessened by assembling the graphite in smaller flakes, as in fine-grained gray iron castings, Fig. F, Plate 7, Fig. E, Plate 12, or even in the spheroids and stars in black-heart malleable castings, Figs. A, B, and C of Plate 4, C of Plate 7; and in Johnson's oxygenated iron, Fig. F of Plate 4.

102. The natural division of the steel-white-cast-iron series of alloys is into the eutectiferous group, which comprises all the white cast irons and may be called "the cast-iron group," and the non-eutectiferous group, which comprises nearly all the carbon steels, and hence may be called roughly "the steel group." This division is indicated above Fig. 10 and in Fig 11. In Fig. 13 the upper line shows the steel group and the lower the cast-iron group. In Plate 3 the middle page shows the microstructure of the steel group and the right-hand page that of the white cast-iron group.

This natural division does not coincide accurately with the industrial division between steel and cast iron, for the members of the eutectiferous group which are poorest in carbon are grouped industrially with the steels because they are usefully malleable. This is shown in the caption above Figs. 10 and 11. Let us hope that the convenience of calling these two groups "the steel group" and "the cast-iron group" may atone for the slight inaccuracy.

103. Resemblance of These Alloys to the Crystalline Rocks.—It is well to impress on our minds at the outset the resemblance of steel and cast iron to the crystalline rocks and other solids which, like them, have formed from the molten or liquid state, and to note that this likeness holds good of their genesis, constitution, mode of aggregation, and even of their defects.

104. The Genesis of iron and steel by the solidification and slow cooling of molten metal resembles that of the crystalline rocks, which have cooled slowly either from the molten state, or at least from a temperature so high that the atoms present in the rock-mass have been free to arrange themselves, and to combine so to form definite compounds, and that these compounds have been free to obey their crystalline laws.

And here we may note that even wrought iron and weld steel (blister steel, etc.), also have formed initially by the solidification of a molten mass, though in their cases the resultant pasty mass has, during and immediately after solidification, by stirring and kneading become mechanically mixed with the cinder or slag of the bath in which the metal has solidified. Nevertheless this stirring and kneading have ceased while the metal is still hot enough to allow its components to obey their crystalline laws, and to form grains which, crude and interfering as they are, yet are true crystals as far as their internal organization is concerned, and have their particles definitely oriented about mathematically definite axes, with true cleavage planes, etc. (§174, and Fig. 26).

105. Constitution.—The rocks of the earth's crust consist of three classes of substances:

1. Simple elements, such as native copper, native gold, graphite, etc.

2. Definite chemical compounds, like feldspar, mica, quartz, hornblende,[1] and

[1] In many of these minerals one base may replace another, manganese replacing iron, etc., and one acid another, but the ratio of acid to base is rigidly fixed. So it is with the corresponding constituents of our alloys. Manganese can replace iron in cementite, yet the ratio of three molecules of metal to one of metalloid remains unchanged.

3. Glass-like obsidians, in which the chemical elements are united, not in any definite ratio, but indeterminately.

Each of these three classes is represented among the constituents of iron and steel, the simple elements by ferrite and graphite; the definite chemical compounds by the iron carbide, Fe_3C, cementite; and the obsidians by the solid solutions such as austenite and probably martensite. These several substances have already been described in §§70, 71, and 95 to 101.

106. Habit of Aggregation.—Just as we find characteristic associations in our rock masses, such as the intermingling of the mica, quartz, and feldspar which so familiarly make up our granites, etc., so we find a like habit of

Fig. 14.—Structure of igneous rock. (Rhyolite from Hot Spring Hills, Pah-Ute Range.) "U. S. Geological Exploration of the Fortieth Parallel, VI, Microscopical Petrography," Plate VIII, Fig. 1.

association of the two most important constituents of iron and steel, ferrite and cementite, in an interstratified mass called pearlite.

Again, as the constituents of rock masses often arrange themselves in a cellular structure, as kernels of one kind of mineral matter within shells of another kind, as shown in Fig. 14, so Fig. B, Plate 1 shows such a common and typical arrangement of kernels of pearlite within enclosing shells of ferrite in steel. In each case we have the shooting out of spines from the shell into the kernel, in well-defined parallelism to fixed crystalline axes.

Here, indeed, the resemblance is extremely close. Our cast irons and very many other alloys consist essentially of a eutectic, or the alloy of lowest solidifying temperature, plus an excess substance, and this relation of eutectic to excess is of the very greatest importance. But here they imitate the igneous and metamorphic rock masses, which very often have this same con-

stitution, in the genesis of which the eutectic and the excess play the same part that they play in the genesis of these alloys. In both cases the excess forms idiomorphic masses, and the eutectic occupies the spaces left between them. The eutectic of cast iron is shown in Figs. F and G of Plate 1. Fig. B of Plate 6 shows this same honeycombed eutectic as a ground mass, in which lie idiomorphic dendritic masses of the excess substance, the austenite, which here forms black bats-wing areas. A rock eutectic is shown with a like banded structure as the ground mass of Fig. E of Plate 6, surrounding the light idiomorphic prismatic crystals of plagioclase. In each case the eutectic itself is a conglomerate made up of distinct particles of radically unlike constituents, of which one is identical with the excess substance, the austenite in our cast iron and the plagioclase in our rock mass. The parallelism is complete.

Fig. 15.—Columnar structure in steel ingot, from the Author's collection, actual size.

107. Columnar Structure.—Again the columnar structure familiar to us in the Palisades of the Hudson, the Giant's Causeway, and like rock-masses, forming enormous columns, is reproduced both in metals and in ice. For instance the columns of the Palisades were formed during the slow cooling of the rock-mass which they form; and they stand upright, *i.e.*, with their length at right angles with their upper surface, which was the cooling surface, the surface through which the heat escaped from them while they were cooling down and changing from a molten glass or obsidian to a solid rock. We find in slowly cooled metals and slags this same columnar structure, with the columns standing at right angles to the cooling surface, *i.e.*, to the outer surface, as is shown in Fig. 15. In large blocks of ice, especially in the ingots of artificial ice which we see about the streets, we can often trace this

columnar structure, with the columns at right angles to the cooling surface, *i.e.*, to the sides of the ice-ingot.

But this geometrical arrangement may well be due to different causes in the two cases, the columnar structure of the solidifying ingots of metal, ice, or slag representing the survival of the crystals normal to the cooling surface in the struggle with those lying at other angles to the surface, essentially because, for given rate of linear growth, the tips of the former reach a region from which solidification is faster. Each such columnar mass is a crystal or group of crystals of like orientation, whereas any one of the basaltic columns, such as those of the Giant's Causeway, is made up of crystals lying at all angles.[1]

FIG. 16.—*A*, superficial blow-holes; *B*, Deep-seated blow-holes; *C*, Pipe.

108. Defects.—Next you find in solidified ingots of steel a contraction-cavity called a "pipe" at the upper end of the axis of the ingot (Fig. 16); and you will generally see a similar pear-shaped cavity in the upper end of the ingots of artificial ice met in the streets.

Next the beautiful specimens of minerals which adorn our mineralogical cabinets generally form in the cavities or "vugs" as they are called in the rocks; beautiful crystals of iron at times occur also in the cavities in our steel ingots as shown in the frontispiece and in Fig. G of Plate 2. In the same way we will often find most beautiful minute crystals of ice in the pear-shaped pipe of ingots of artificial ice.

Again, just as the distortion of the rocks of the earth's crust causes what is called a "schistose" structure and sometimes crumples the strata, so we find like schistosity and crumpling in metals which have undergone plastic deformation. For instance, such a drawn out schistose structure is shown in Fig. G of Plate 7, in which vertical white bands of ferrite are seen, in part bordering a black band of graphite. Such bands are called "ghosts." Again the lamellæ of pearlite in Fig. F of Plate 2, here shown with their general direction nearly ver-

[1] Some believe that this columnar basaltic structure is of heat-convectional origin, each column representing a convection current cell in the solidifying mass, the current rising in the axial part of the cell and descending in its peripheral region. It is true that, in the case of thin layers of liquids, this convectional origin of the cell structure is well established, and that the basaltic columns are in general so nearly vertical that we may believe that they were strictly vertical originally, a condition necessary to the formation of heat-convection cells. But there is a certain difficulty here. It is easy to understand how convection currents should cause short stumpy cells, but the difficulty of referring these cells to convection alone increases both with the viscosity of the liquid and with the length of the cells, not indeed with their absolute length but with the ratio of their length to their width, in short with their departure from the stumpy form. Both these difficulties lie in the way of explaining as convection cells the very long narrow columns which form in so viscous a substance as basalt. Beyond this, they are normal to the cooling surface even in cases in which this was clearly much inclined at the time of their solidification (C. H. Desch, private communication, Oct. 24, 1914). An origin like that of the drying cracks in clay suggests itself. See Desch, "Solidification of Metals from the Liquid State," *Journ. Inst. Metals*, 1914, vol. 9, No. 1, p. 113. See also §405 and §§772 *et seq.*

tical, have been much crumpled by the punching which this piece of steel has undergone. Such lamellæ were initially only gently curved in other parts of this specimen. But note the crumpling of the right-hand ones into a W, and the bending of the lamellæ of cementite near the left-hand end of the photograph into a V.

Other cases of crumpling and shattering are shown in Figs. A to D, of Plate 26.

Fig. C shows the great distortion of the slag in a piece of muck bar, or crude wrought iron, bent double and hammered down close while at a white heat. It recalls very closely the foldings in the crystalline rocks. The white islands lying tandem in Fig. A are the fragments of what were long atolls of cementite, left by the divorcing or disentwining of the pearlite in a steel of 0.21 per cent. of carbon on slow cooling and consequent long exposure to a temperature shortly below the transformation range, about 650°. This shattering was brought about in the tensile test when cold.[1]

The long vertical strings of white vertebræ in Fig. B are the fragments of the great white cementite islands in a steel of 2.04 per cent. of carbon, shown in their unbroken state in Fig. I2 of Plate 3. In this case the shattering was brought about by bending a $\frac{5}{16}$-in. square bar of this steel by means of gentle blows while at about 700°.

Finally, gases evolved during solidification cause gas-bubbles or "blow-holes" in ingots of ice and of steel, and also in glass (A and B, Fig. 16). These blowholes form at a time when the mass is still fluid enough to be pushed aside by the particles of gas evolved within it, so that these come together to form gas-bubbles; yet not fluid enough to permit these bubbles to rise by gravity to the upper surface and thus escape. Hence these bubbles remain entangled in the viscous mass.

[1] See Howe and Levy, *Trans. Amer. Inst. Mining Eng.*, 1914, vol. 50, p. 532.

CHAPTER 6

OUTLINE OF THE CONSTITUTION OF IRON (CONCLUDED)

The Cast-iron Series

109. Summary.—The members of the cast-iron series differ from each other primarily in their content of graphite. This series may be regarded as an offshoot from the steel-white-cast-iron series, and as equivalent to that series plus graphite. This relation of the two series is shown in this chapter both by means of the pig irons (§§110 to 115) and by means of the industrial foundry products (§116) and both as regards the content of total carbon and of graphite on one hand, and as regards the content of total carbon and of combined carbon on the other hand. The distinction between the various grades of pig iron differs radically from that between the various classes of steel, for whereas these steels differ from each other primarily and greatly in their total carbon content and are alike in being practically graphiteless, the grades of pig iron, though differing from each other relatively little in their content of total carbon, differ primarily and greatly in the degree to which that carbon is graphitized (§113).

In constitution (§117) the various members of this cast-iron series are conglomerates, each consisting essentially of two very unlike parts. One of these parts is a metallic matrix, identical in nature with some member or members of the steel-white-cast-iron series, that is with some steel or some white cast iron, and the other is graphite, which as a foreign body weakens and embrittles that matrix. Many illustrations of this constitution of the cast irons are given (§§119 and 128), and in particular several micrographs make this constitution clear to the eye (§118).

The various members of the cast-iron series differ from each other in the degree to which their metallic matrix is weakened and embrittled by their content of graphite, a degree which increases roughly with their graphite content (§§119 and 128), and with the degree to which that graphite is spread out into thin skeleton-forming flakes instead of being gathered into compact shapes (§118); and they further differ from each other in the carbon content of their metallic matrix, and hence in the properties of that matrix. In this latter respect, which is of very great importance for the foundry products, the basis of distinction between these various cast irons is identical with that between the various industrial classes of carbon steel (§127).

Confining ourselves in this chapter to the constitution of this cast-iron series, we will take up in Chapter 11 the genesis of its characteristic member, its graphite.

110. The relation of the gray cast iron series to the steel-white-cast-iron series is illustrated by Figs. 17 to 19, which are modified from Fig. 10 to this

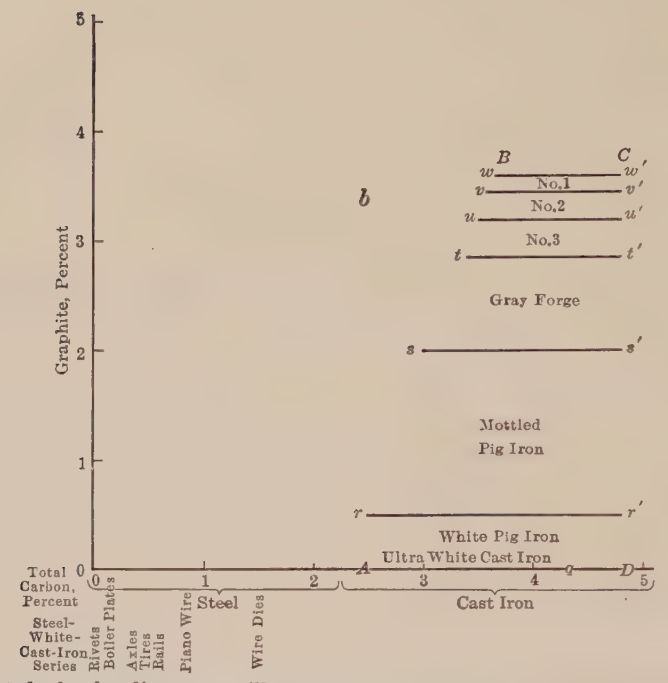


FIG. 17.—First draft of a diagram to illustrate the relation of the gray pig irons to the steel-white-cast-iron series, as regards their content of total carbon and of graphite.

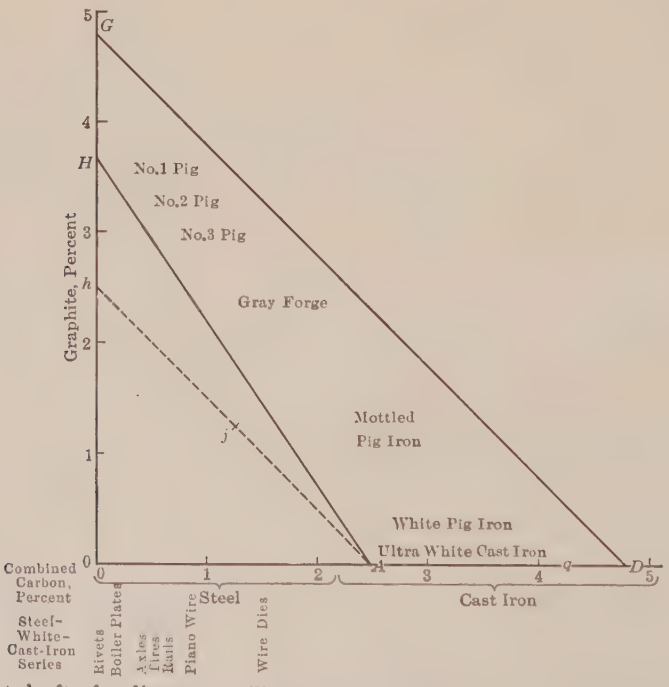


FIG. 18.—First draft of a diagram to illustrate the relation of the gray pig irons to the steel-white-cast-iron series, as regards their content of combined carbon and of graphite.

end, by giving it a vertical dimension, to represent the percentage of graphite present. Fig. 10 showed us that the white cast irons are simply a prolongation of the series of steels, with progressively increasing content of carbon, nominally all combined as cementite. Fig. 17 adds to this the conception that the gray cast irons differ from the white ones essentially in having more or less of their carbon present as graphite instead of as cementite. They may be regarded as white cast irons graphitized. This graphitization is usually far from complete, though in some exceptional cases, as in black heart malleable castings, it may be practically complete. For our present purpose

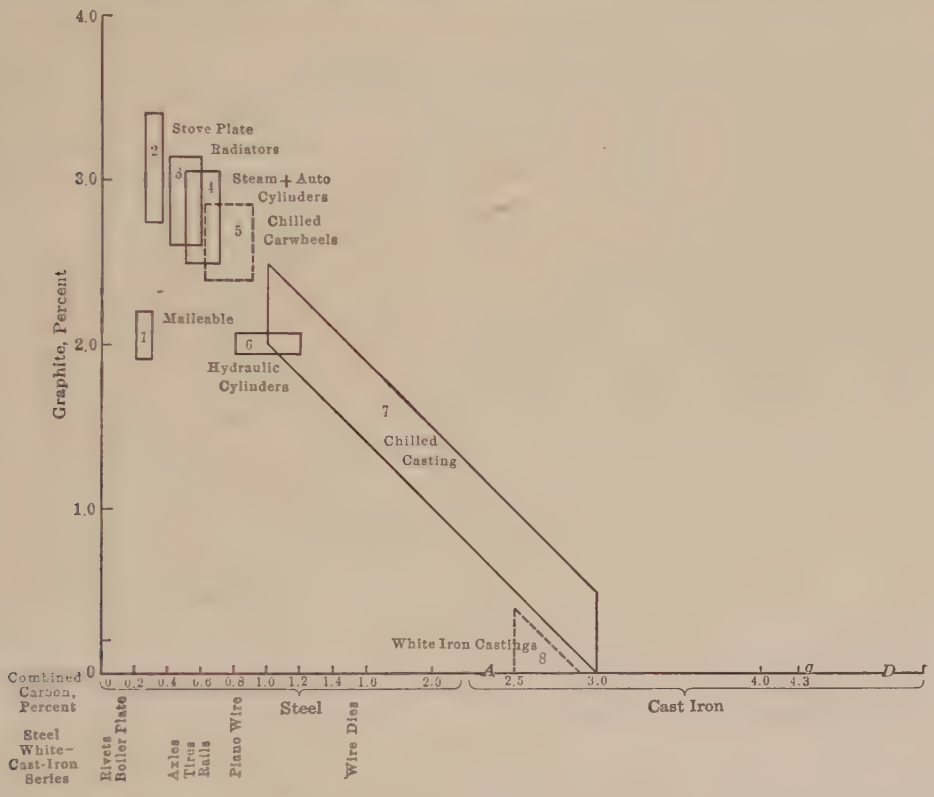


FIG. 19.—First draft of a diagram to illustrate the relation of the cast iron foundry products to the steel-white-cast-iron series, as regards their content of combined carbon and of graphite.

these malleable castings may be regarded as simply one member of the gray cast-iron series, though a member with distinctive valuable properties.

This diagram gives us a rational way of taking in at a glance the various cast irons and the various steels as parts of a single series, the gray cast iron part of the series forming an offshoot, $ABCD$, from the steel-white-cast-iron part, AD. Some may prefer to look at them as two connecting series, connecting through the white cast irons, the steel-white-cast-iron series running horizontally in Fig. 17 from A to D, and the cast-iron series standing there in column. But this is a difference rather in words than in thoughts.

Again we may divide the whole series mentally at the point A, into the

steels on one hand, and the cast irons, white and gray together, on the other hand. From this point of view there is the striking difference between the left-hand or steel part of the series in Fig. 17 and the right-hand or cast-iron part, that whereas the carbon of industrial steels exists wholly in the state of combination as cementite in slowly cooled or annealed steels, and as martensite or austenite in most of the rapidly cooled or hardened ones, the carbon of the cast iron may be distributed between the states of cementite[1] and graphite in any ratio, being wholly combined in ultra white cast iron and wholly graphitized in ultra gray iron. In other words, in the cast irons there may be any degree of graphitization.

111. Additional Graphical Representation of this Relation.—The source and measure of the most valuable qualities are, in the last analysis, the same for the cast iron foundry products as for steel, the content of combined carbon or of cementite. But before trying to grasp this conception let us familiarize ourselves with the conditions by means of the three diagrams, Figs. 17 to 19, which show the relation of these two series to each other in two different ways, and as applied to two different kinds of objects. In following the first of these ways we consider the content of total carbon and of graphite; in following the second we consider the content of combined carbon and of graphite. The two series of objects studied are first the various vague classes of pig iron, and second the industrial foundry castings, slightly less vague.

In all three of these diagrams the ordinates are the same, graphite content. Hence in all three the steels and white cast irons, because nominally free from graphite,[2] are represented by a horizontal line, OAD, which is the same in all, whereas the column which represents the gray cast irons is vertical or inclined according to the point of view.

It is because the steels are in general nominally free from graphite that their names are written below the line OA in Figs. 17 to 19, lest writing them above that line should suggest a certain degree of graphitization.

112. This Relation Illustrated by the Pig Irons. Total Carbon Content and Graphite Content.—In Fig. 17 the relation of the various grades of pig iron to the steel-white-cast-iron series is shown roughly. If, in comparing these various grades as made under any one set of conditions which are constant, except in so far as they affect the grade itself, we pass from the white irons through the mottled to the grayest and coarsest of all here shown, and if we take the average of a large enough number of cases, we find that these various grades differ from each other essentially, first, in that they contain

[1] It may be that, on their removal from the mould, some chilled cast iron castings, especially those rich in manganese, have an appreciable amount of martensite, and hence of carbon combined in this martensitic state and not simply as cementite. But in the annealing of these castings this martensitic combined carbon shifts over into cementite.

[2] Hyper-eutectoid steels have indeed a tendency to graphitize, which is the stronger the more strongly hyper-eutectoid they are. For instance, in treating file steel of 1.25 to 1.50 per cent. of carbon the procedure must be regulated with care to prevent graphitization. But this graphitizing tendency is in fact restrained, so that the actual conditions approach pretty closely to the ideal ones here set forth, absence of graphitization in the steels and white cast iron, and much graphitization in the gray cast irons.

successively more and more graphite, and second, in that they contain successively more and more total carbon. The first of these differences leads us to set them successively higher and higher, and the second leads us to incline to the right the column in which they stand, as has been done in placing their names in Fig. 17.

What we should here recognize is that these various pig irons differ from white pig iron primarily in the degree in which their carbon exists as graphite, and hence that they are represented by an area or column resting on the white cast irons as a line, which is the prolongation of the steels.

In their basis of classification, graphite content, they thus differ radically from the industrial steels. These are classified by their content of combined carbon, which is sensibly the same as that of their total carbon content. Each industrial class of steel has in a sense a nearly fixed carbon content.[1] The various classes of steel differ from each other not only primarily but greatly in carbon content, incomparably more than the various grades of pig iron differ in this respect. For instance, there is about thirty-five times as much carbon in certain kinds of die steel as in the steels poorest in carbon.

113. Each Grade of Pig Iron Varies in Graphite Content.—Under varying conditions, varying content of sulphur, silicon, and manganese, size of pig, kind of mould, etc., the graphite content of each of these grades of pig iron varies so greatly that some very gray irons may contain less graphite than some white irons. Thus Bell gives a case in which a white Clarence pig contained 2.209 per cent. of graphite, whereas a No. 1 Clarence pig, which normally should be very graphitic, contained only 2.07 per cent.[2] Moreover, the total carbon content of each grade also varies with the conditions within the blast furnace. Hence this arrangement of names shown in Fig. 17 must not be taken as quantitatively true except for some one set of conditions, but rather as a first approximation made for the purpose of illustrating the matter in hand, this relation of these pig irons to the steel-white-cast-iron series.

The grade names written in Fig. 17 should be taken as a first approximation to the center of gravity of their several grades. Strictly speaking, the areas of these various grades should not only overlap, but overlap widely. The areas here drawn without this overlapping must therefore be taken as representing a rough estimate of the most densely settled area for each grade. The case is somewhat like that of the Scots who live across the English border and *vice versa*.

If, in view of these variations of the content both of total carbon and of graphite in each grade, we try to represent in such a diagram, not the pig irons made under some one set of conditions, but generically all pig irons

[1] The total carbon content of steel for a given purpose is rather closely fixed, in a sense which can well be made clear by an example. While it is true that some engineers would have only about 0.35 per cent. of carbon in their rails, and that others would have 0.70 per cent. in certain classes of rails, yet each of these engineers will tolerate only a narrow variation in the carbon content of rails of a given size and for a given service. And so on with the various important uses.

[2] "Principles of the Manufacture of Iron and Steel," Routledge, Spon, London, 1884, p. 158.

which could be made under any probable conditions, we should represent the various grades by a series of horizontal bands, as sketched by the lines *rr'* to *ww'* in Fig. 17. These limits, too, should be taken as a first approximation, because their position, too, varies with the attendant conditions, and also because it is not known that they would be strictly horizontal under any given series of conditions. I have drawn them horizontal in this first approximation, because the grade of a pig iron, if judged by its fracture, seems to depend primarily on its graphite content, at least for given coarseness of the graphite flakes themselves.

So, too, because there is no agreement as to the boundary between steel and white cast iron, and because there is no natural upper limit to the carbon content of pig irons, I have drawn no right- or left-hand boundaries of these horizontal areas representing the various grades of pig iron; and though the designating letters *r*, *r'*, etc., must needs have some position, that which they have here should be taken as an approximation.

The left-hand imaginary boundary, *AB*, is strongly inclined, because the possible graphite content must needs increase with the content of total carbon. If there is but 2.50 per cent. of total carbon there cannot possibly be more than 2.50 per cent. of graphite; whereas if there is 4.50 per cent. of total carbon, there might be 4.50 per cent. of graphite, for the whole of this 4.50 per cent. might be graphitized.

For a like reason the right-hand boundary also is in part inclined. As explained in §307, pure carbon-iron industrial pig iron is not likely to contain more than 4.30 per cent. of combined carbon, and indeed it may necessarily be limited to this content. This limits the total carbon content of the industrial ultra white or ideal white pig irons to 4.30 per cent., because that total carbon is identical with their combined carbon.[1] Note that this

[1] Though pure carbon cast irons are unlikely to contain more than 4.30 per cent. of combined carbon, those rich in manganese, as for instance the spiegeleisens, may contain much more. Bell gives the greatest content of carbon found in the records of the Clarence works, and also an extremely small content of carbon, as follows:

Table 10.—Cast Irons Unusually Rich and Unusually Poor in Carbon

	No. 1 pig iron	No. 3 pig iron	No. 1 pig iron	
			Original pig	Re-melted for castings
Carbon	4.50	4.82	1.61	1.65
Silicon	1.28	1.58	0.98	0.61
Manganese	0.72	0.62	0.12	0.47
Phosphorus	1.39	1.35	1.37	1.25
Sulphur	0.08	Slight trace	0.16	0.29
Titanium	0.10	0.22	....	....
Calcium	.....	0.23	0.40	0.81
Magnesium	.....	0.05	0.29	0.43
Iron	92.47	92.00	95.15	94.42
Total	100.54	100.87	100.08	99.93

(I. L. Bell, "Principles of the Manufacture of Iron and Steel," Routledge, Spon, London, 1884, p. 157.)

identity is confined to ideal white irons, which by definition are free from graphite.

The carbon content of the gray irons is not thus limited because, though they cannot contain more than 4.30 per cent. of combined carbon, they may in addition contain much graphite. Hence for our present crude approximation we may regard the right-hand boundary as running obliquely from q to r', and thence vertically to w'.

114. This Relation Illustrated by the Pig Irons. Content of Combined Carbon and of Graphite.—If next, in Fig. 18, we represent these same grades with graphite content as ordinate as before, but substituting content of combined carbon for that of total carbon as abscissa, we thereby incline the column strongly to the left. Of course in passing from Fig. 17 to Fig. 18 the steels and the white irons do not shift their position, because they have the same content of combined as of total carbon, and hence stand in the same position when abscissa is combined carbon as when it is total carbon. But because mottled pig iron has some of its carbon as graphite, it has just so much the less combined carbon, and hence its name must be shifted to the left far enough to represent the difference between its combined carbon content here shown, and its total carbon content shown in Fig. 17. Because gray forge pig has still more graphite, its shifting to the left must be still greater, and in brief as the graphite content increases from grade to grade, so must the names of the successive grades be moved successively farther to the left.

What has been said of these grade names applies equally to the areas representing the various grades of pig iron, shown in Fig. 17 by the horizontal areas between the lines tt' and ww'.

Precisely the same reasoning that has led us to move any given name to the left applies to any given point in what were the undrawn boundaries of these areas.

Because the presence of n per cent. of graphite makes the combined carbon content n per cent. less than the total carbon content, it follows that, in passing from Fig. 17 to Fig. 18, and substituting combined for total carbon content as abscissa, the distance which each point of Fig. 17 must be moved to the left, to represent this n per cent. of decrease in its abscissa, is identical with the n per cent. ordinate of that point. But this means swinging the lines AB and DC approximately 45° to the left in passing from Fig. 18 to Fig. 19, as has been done.

115. A Missing Area.—Now that we have seen that the steels are represented in Fig. 18 by the line OA, the white cast irons by the line AD, and the cast irons as a whole, white and gray, by the area $ADGH$, we naturally ask what the triangle OAH represents. The part OAh represents what the steels would be if they were graphitized. For instance, the limiting steel, A, if it were graphitized completely, would be represented by the point h, if half graphitized by the point j, etc. Because the graphitization of steels is almost wholly restrained industrially, this triangle has no industrial exist-

ence. But for that matter a careful study might enable us to assert that certain parts of the trapezoid *ADGh* have no industrial existence.

116. The Relation of the Gray Cast-iron Series to the Steel-white-Cast-iron Series Illustrated by the Iron Foundry Products.—If we now turn from this broad classification made to include all reasonably probable pig irons, and try to represent instead certain of the more important classes of foundry products in Fig. 19 on the plan used in Fig. 17, we get the figures here shown.

Hydraulic cylinders, for instance, may contain from 2.50 to 3.50 per cent. of total carbon, and of this anywhere from 0.80 to 1.20 per cent. may be present in the combined state. The four corners of the rectangle which represents them are accordingly as follows.

TABLE 11.—POSITION OF THE CORNERS OF THE AREA REPRESENTING CAST-IRON HYDRAULIC CYLINDERS

	Combined or cementite carbon	Graphite	Total carbon
A	0.80	1.70	2.50
B[a]	0.80	2.30	3.10
C	1.20	1.70	2.90
D	1.20	2.30	3.50

[a] The method of finding the point B should be explained. It might be thought that for this point we should assume the lower limit of combined carbon, 0.80 per cent., with the higher limit of total carbon, 3.50 per cent., which would imply a graphite content of 2.70 per cent. But it is improbable that any given specimen would have both the lowest combined carbon content and also the highest total carbon content, because this would imply an unwelcomely large content of graphite, which might lead to leaking. Hence it is more reasonable to take for point B the graphite content of point D, due to the coinciding of the highest content of combined and of total carbon.

A like consideration applied to the point C gives us a graphite content of 1.70, or that resulting from the coinciding of the minimum content of total and combined carbon. But these considerations hardly apply to the chilled castings and white iron castings, which are therefore given the full limits.

So with the other kinds of castings represented in Fig. 19. The width of the several figures shows the range of their total carbon content, and their height the range of graphite content.

Here we should recognize important respects in which these various kinds of foundry products differ on one hand from the corresponding steel products, and on the other hand from the various grades of pig iron, white mottled, etc. Each of these foundry products, taken as a group, differs from the various steel products, not only in having a very much wider variation in carbon content, but also in having a wide variation in graphite content. On the other hand each of them differs from the grades of pig iron, in having a much narrower range both of total and of combined carbon content.

Hence the figures which in Fig. 19 represent these foundry products differ from the fragments of the line *OA* which represents the various industrial classes of steel, both in having greater horizontal length, and in having a vertical dimension; and they differ from the imaginary areas which represent

the broad classes of cast iron, mottled, close gray, etc., in being much shorter horizontally.

TABLE 12.—FIRST APPROXIMATION TO LIMITS OF CARBON CONTENT FOR VARIOUS CAST-IRON OBJECTS

No. in Fig. 19	Name	Combined carbon, per cent.	Graphite, per cent.	Total carbon, per cent.
1	Malleable castings	0.20 to 0.30	1.90 to 2.20	2.10 to 2.50
2	Stove plate	0.25 to 0.35	2.75 to 3.40	3.0 to 3.75
3	Radiators	0.40 to 0.60	2.60 to 3.15	3.0 to 3.75
4	Steam and auto-car cylinders	0.50 to 0.70	2.50 to 3.05	3.0 to 3.75
5	Chilled car wheels and rolls	0.60 to 0.90	2.40 to 2.85	3.0 to 3.75
6	Hydraulic cylinders	0.80 to 1.20	1.95 to 2.05	2.75 to 3.25
7	Chilled castings	1.0 to 3.0	2.0 to 0.50	3.0 to 3.50
8	White-iron castings	2.50 to 2.90	0.0 to 0.40	2.5 to 2.90

The area representing chilled castings in Fig. 19 is an exception because of the great variation in the degree of graphitization of different parts of one and the same casting. The combined carbon content of the chilled face may be very great, though that of the gray interior may be relatively low.

The extreme variations in the content of total carbon and of graphite which actually occur in these various kinds of foundry cast-iron products, in the present astonishingly crude state of the art,[1] are indeed much greater than those which I have adopted for the purposes of illustration in Fig. 19. Hence these figures should be taken only as a rough surmise of the limits within which certain of the more intelligent founders aim to have their products lie.

Many founders give a smaller content of combined carbon than I have called for in order to avoid the complaints from the machine shop that their castings are too hard to machine, thus sacrificing strength to softness, often unwittingly we may suspect.

117. Gray Cast Irons a Conglomerate of a Metallic Matrix with a Graphite Skeleton.—Turning to the name "No. 1 pig" in Fig. 18, and noting that it stands vertically above the hypo-eutectoid steels, we are brought face to face with the important conception that this iron is the equivalent of this steel as regards its content of combined carbon. From this it follows that it is also the equivalent of such steel as regards the kind and proportion of its microscopic constituents. Let us test this step by step.

[1] For a very valuable collection of analyses of industrial cast-iron castings see *Report of the Committee on Standards for Iron Castings* by J. J. Porter (*American Foundrymen's Association*, June, 1910). Here in fact we find the most extraordinary variations in the total and in the combined carbon content of important castings, such as automobile cylinders, made for one of the most exacting and intelligent classes of consumers. The total carbon content of these varies from 2.52 per cent. to 4.16 per cent., and their combined-carbon content varies from 0.03 per cent., or less than that of much of the best rivet steel, to 0.90 carbon, or much more than that of most of the strongest rope wire. Compare also C. A. Meissner, *Iron Age*, 1888, vol. 41, p. 919; abstracted in *Journ. Iron and Steel Inst.*, 1888, No. II, p. 318.

This combined carbon is scattered throughout the metal in the form of minute particles of cementite, as it is in the case of steel. And exactly as in the case of steel this cementite contains fifteen times as much iron as carbon, according to its formula Fe_3C. Moreover, exactly as in steel, all the remainder of the iron not contained in this cementite exists in the free state, that is as ferrite. Moreover this cementite and this ferrite are associated exactly as in steel, in the mechanical mixture called pearlite, in the ratio of 1:6.4. Any ferrite or cementite present in excess of this ratio exists in the pro-eutectoid or free state, as pro-eutectoid ferrite or pro-eutectoid cementite.

Passing down our inclined column in Fig. 18 to mottled pig, which stands vertically above the relatively low-carbon hypo-eutectoid white cast irons, we find that apart from its containing less graphite, its constitution differs from that just described essentially in containing some eutectic in addition to pearlite and pro-eutectoid cementite. Moreover, the pearlite and the eutectic have the characteristic outlines of the primaustenoid and the eutectic of hypo-eutectic white cast irons. Thus far, then, this cast iron apart from its graphite content, is identical in constitution with a hypo-eutectic white cast iron, in that it consists of primaustenoid and eutectic, each of which is identical in constitution and in outline with the primaustenoid and the eutectic of hypo-eutectic cast iron.

So with every other point in the inclined column *ADGH* of Fig. 18. It contains the same constituents in the same proportions and in much the same topographical arrangement as the member of the steel-white-cast-iron series above which it stands vertically. It contains these and something else.

Up to this point these gray irons are thus identical in constitution with the various members of the steel-white-cast-iron series. But in coming up to this point we have disposed of all the combined carbon and all the iron. What remains? Apart from the silicon, phosphorus, sulphur, manganese, etc., which are present in steel, in white cast iron, and in gray iron, and hence constitute no difference between these classes,[1] there remains no constituent to distinguish the steel-white-cast-iron series from the gray cast irons except the graphite of the latter.

The cementite and the ferrite, which thus represent the combined carbon and the iron of our gray cast iron, form here, exactly in the steels and white cast irons, a distinctly metallic mass, as typically metallic as

[1] The silicon and phosphorus and perhaps part of the manganese are dissolved in the ferrite, and the remainder of the manganese displaces part of the iron in the cementite and in the iron sulphide. Cementite, indeed, should have the formula $(FeMn)_3C$, implying that iron and manganese replace each other in indefinite ratios. Hence these various foreign elements modify the chemical composition of these constituents, without adding any new constituent. In particular the silicon content of cast iron is habitually greater than that of steel, so that the ferrite of cast iron is habitually richer in silicon than that of steel. Nor does the sulphur of cast iron add a new constituent, for the sulphides of iron and of manganese which it forms are the same in character as in steel, and substantially the same in quantity for our present purpose. An exception might indeed be made of the phosphorus compounds present in pig iron for the basic bessemer process. But our present broad survey of the iron-carbon alloys need not be complicated here by special consideration of these foreign bodies, foreign at least to our present point of view.

anything which exists. But graphite is as non-metallic in cast iron as in a lead pencil. It is scattered, usually in the form of flakes, through the metallic mass which represents the combined carbon and the iron jointly, that metallic mass of pearlite with ferrite or cementite or austenoid, as shown in Fig. 11. Hence we may reasonably say that a gray cast iron is a conglomerate consisting of a metallic matrix and a graphite skeleton. Moreover this matrix in any given gray iron is approximately equivalent to some member of the steel-white-cast-iron series, and for any given specimen that member is clearly the one vertically below the name of the cast iron in Fig. 18, if this is placed correctly.[1]

But because the degree of graphitization varies greatly even in adjoining fields in a given specimen of cast iron, this matrix in fact corresponds in different parts of the specimen to different members of the steel-white cast-iron series. Examples of this are given in the next section.

In short, each member of the gray cast-iron series consists of a metallic matrix, approximately equivalent, on an average of its different parts, to that member of the steel-white-cast-iron series to which it corresponds in content of combined carbon, with its continuity broken up more or less completely by masses of graphite as by a foreign body. A No. 1 pig iron may represent a low-carbon steel matrix much broken up by abundant flakes of graphite; a No. 2 iron a higher carbon steel matrix less broken up by such graphite. And in general as we pass down the inclined column of Fig. 18 the succeeding grades of pig iron are conglomerates of which the metallic matrix is equivalent to successively higher-carbon members of the steel-white-cast-iron series, broken up by successively smaller quantities of graphite, till when we reach the limit of the gray cast iron series we find that it is white cast iron, AD, of Fig. 18, and thus is identical with the right-hand part of the steel-white-cast-iron series.

In this progress we should at first expect the matrix to reach its greatest strength when it contains about 1 per cent. of combined carbon, from the fact that the tensile strength line of Fig. 10 here reaches its maximum. Hence we should expect that, for given graphite content, the strongest iron would be one with about 1 per cent. of combined carbon. Further reflection inclines us to estimate a rather smaller content of combined carbon content for this strongest cast iron. The reason is simple. The eutectic itself is probably a source of weakness, at least when there is enough of it to form extended masses, because its perfectly brittle cementite plates

[1] See the Author, Discussion of Kreuzpointner's paper on "The Chemistry and Physics of Cast Iron," *Journ. Franklin Inst.*, 1900, vol. 149. Though many others had probably conceived this relation between the steels and the cast irons, it was here enunciated for the first time so far as I know. It was received with great incredulity.

"The Influence of Silicon and Sulphur on the Condition of Carbon in Iron," *Trans. Amer. Inst. Mining Eng.*, 1900, vol. 30, p. 719.

"The Constitution of Cast Iron," *ibid*, 1901, vol. 31, p. 318, and discussion, p. 995.

"On the Constitution of Cast Iron," *Proc. Amer. Soc. Testing Materials*, 1902, vol. 2, p. 246.

"Iron, Steel, and other Alloys," Boston, 1906, Second edition, p. 431.

are so thick that any one of them in breaking is likely to start a crack which will lead to the rupture of the whole piece.

In order that the mass shall be free from this defect it ought to be nearly free from eutectic. But as all cast iron is eutectiferous at the moment of solidification, this means that the graphitizing conditions, silicon-content included, should be strong enough if not to graphitize the whole of the eutectic, at least to leave only small fragments of it ungraphitized. But the combined carbon content of the matrix varies greatly in one and the same casting. A cast iron with as much as 1 per cent. of combined carbon may have more than 1.70 per cent. in certain parts. Such parts would be eutectiferous, and hence weak, according to this line of reasoning. In order to avoid the presence of eutectic, or at least to avoid any harmfully large masses of eutectic, it may indeed be necessary to reduce the average combined carbon content of the mass well below 1 per cent. In such reduction, the gain of strength due to avoiding the presence of eutectic may well outweigh the simultaneous loss of strength which we might expect from making the matrix, on an average of its whole, correspond to a lower carbon and hence weaker steel. Here we must remember as ever that it is the weakest link that determines the strength of the chain, and thus that it is the quantity of eutectic in the most eutectiferous parts of the specimen that limits the strength. Such local weakness is not helped by great strength in other parts caused by their matrix representing a strong 1 per cent. carbon steel.

If I am right in this, the reasonable way to make a strong cast iron is to bring about thorough graphitization of the eutectic. To prevent this graphitization from causing too wide plates of graphite, it should occur late during solidification, or better soon after solidification. This can be brought about only by means of a rather large silicon content. In order to prevent the presence of the resultant large quantity of graphite from inducing almost complete graphitization of the cementite liberated at lower temperatures, the cooling from say 800° down should be rapid enough to restrain graphitization. In some cases this combination can be brought about by removing the casting from the sand when it has cooled to 800°, and thenceforth cooling it in the air. This method lends itself to cases in which great numbers of like castings are made, for here the exact length of time that the casting is to stay in the sand can be determined. In other cases it may be better to make the initial cooling complete and relatively rapid, to reheat the casting to 800°, and then to cool it rapidly enough to restrain graphitization, yet slowly enough to allow the transformation from austenite to pearlite to complete itself.

To sum this up, our natural working hypothesis, which must be controlled by careful and extended observation both chemical and microscopic, is that the greatest strength should be had: (1) by having the graphite masses as compact as possible; (2) by having the content of combined carbon large, say between 0.70 and 1 per cent.; and (3) by having the total carbon content,

TABLE 13.—THE WEAKENING AND EMBRITTLING ACTION OF GRAPHITE, AS SHOWN BY THE GREATER STRENGTH AND DUCTILITY OF MALLABLE CASTINGS THAN OF GRAY CASTINGS, AND OF STEEL CASTINGS THAN OF EITHER

Number	Source and description	Mechanical properties: Tensile strength, lb. per sq. in	Mechanical properties: Elongation per cent.	Chemical composition, per cent.: Carbon: Combined	Carbon: Graphite	Carbon: Total	Si	Mn	P	S	V	Cu
	Cast-iron castings. Standard properties											
1	*American Society for Testing Materials Specification*, 1914. Gray iron, less than ½ in. thick, light.	Not less than 18,000	.....	.....	.....	.....	.....	...no	t over	0.08	.....	.....
2	*American Society for Testing Materials Specification*, 1914. Gray iron, medium weight..	Not less than 21,000	.....	.....	.....	.....	.....	...no	t over	0.10	.....	.....
3	*American Society for Testing Materials Specification*, 1914. Gray iron, not less than 2 in. thick.	Not less than 24,000	.....	.....	.....	.....	.....	...no	t over	0.12	.....	.....
4	*W. J. Keep.* Castings for compressor cylinders, valves, high pressure work, etc.	22,000–25.000	.....	.....	.....	.....	1.20–1.50	0.50–0.80	0.35–0.60	under 0.09	.....	.....
5	" " Castings for low pressure cylinders, gears and pinions, etc.	20,000–23,000	.....	.....	.....	.....	1.50–2.00	0.50–0.80	0.35–0.60	under 0.08	.....	.....
6	" " General car and railway castings, pulleys, small castings, and agricultural work.	18,000–20,000	.....	.....	.....	.....	2.20–2.80	under 0.70	under 0.70	under 0.085	.....	.....
7	*W. A. Jenkins.* 11 American (U. S.) and I French auto. cylinders.	14,710–25,260	.....	0.028–0.75	2.59–3.29	2.87–3.91	1.60–3.56	0.22–0.823	0.05–0.820	0.053–0.159	.....	.....
8	Specification for mortars made from charcoal gun iron.	30,000–37,000	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....
	C. Jüngst. Original dimensions of test-bars: Square or round / Width, inches / Length, inches											
9	Best..... Square....	1.18	39.37	28,400	.....	.....	.....	.....	.....	.....	.....	.....
10	Best..... Round....	0.79	15.75	35,300	.....	.....	.....	.....	.....	.....	.....	.....
11	Medium.. Square....	1.18	39.37	22,200	.....	.....	.....	.....	.....	.....	.....	.....
12	Medium.. Round....	0.79	15.75	30,300	.....	.....	.....	.....	.....	.....	.....	.....
13	Structural and pipe. Square....	1.18	39.37	17,900	.....	.....	.....	.....	.....	.....	.....	.....
14	Structural and pipe. Round....	1.18	23.62	21,600	.....	.....	.....	.....	.....	.....	.....	.....
	Limits, lb. per sq. in	14,710–35,300	.....	.....	.....	.....	1.20–3.56	0.22–0.823	0.50–0.820	0.053–0.159	.....	.....
	Limits, kg. per sq. mm	10.3–24.8	.....	.....	.....	.....					.....	.....
	Cast-iron castings. Unusually good properties											
15	*M. M'Dowell.* Casting for government work....	49,900	.....	0.93	2.19	3.12	1.34	1.0	0.08	0.03	.....	.....
16	*M. White.* Gun-iron bar cut from 8 in. solid shot.	37,500	.....	.....	.....	2.39	1.27	0.57	0.057	0.044	.....	.....
17	*G. L. Norris.* Ladle test on melt for loco. cylinder.	34,560	.....	0.79	2.34	3.13	1.83	0.43	0.508	0.102	nil	.....
18	*G. L. Norris.* Ladle test on melt for loco. cylinder.	35,600	.....	0.71	2.48	3.19	2.21	0.424	0.546	0.071	0.056	.....
19	*Kent.* Pig iron from Salisbury ores made at Wassaic and Millerton, N. Y.	42,281	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....
20	" Muirkirk, Md., No. 4 iron, tested at Washington Navy Yard.	41,329	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....
21	R. Moldenke.....	36,860	.....	0.51	2.44	2.95	1.83	0.55	0.610	0.100	.....	.....
22	" "	34,010	0.7 in 2"	1.14	1.60	2.74	1.14	0.42	0.360	0.05	.....	.....
23	*G. R. Johnson*, Embreville, Tenn., very close gray.	35,950	.....	0.92	2.84	3.76	1.29	.....	0.179	0.053	.....	.....
24	" " " " " close gray....	35,650	.....	0.97	2.96	3.93	1.23	.....	0.202	0.141	.....	.....
25	*A. L. Colby.* Machine cast pig.....	41,000	.....	0.92	2.46	3.38	2.99	0.95	0.773	0.041	.....	.....
26	*E. Gridley.* Wassaic pig iron, 1883.....	41,349	.....	0.78	2.31	3.09	1.307	1.512	0.294	0.086	.....	.....
27	*A. W. Whitney.* Cast 1–1/4 in. sq., turned 1.128 in. diam.	44,970	.....	0.86	2.12	2.98	1.662	0.488	0.512	0.079	.....	0.160
28	" " " Cast 2 in. sq., turned 1.128 in. diam.	40,070	.....	0.42	2.40	2.82	1.234	0.766	0.416	0.142	.....	0.156
29	" " " Gun iron.....	48,360	.....	0.78	2.31	3.09	1.31	1.50	0.29	0.09	.....	.....
30	*Pittsburg Testing Laboratory.* Mesta Machine Co.	37,500	.....	0.61	2.50	3.01	1.70	0.32	0.530	0.057	.....	.....
31	" " " " " "	36,400	.....	0.59	2.35	2.94	1.70	0.35	0.532	0.046	.....	.....
32	*J. E. Howard, Tests of Metals.* 1893, p. 282. From Poole...	35,780	.....	1.276	2.004	3.280	1.062	0.520	0.520	0.180	.....	0.008
33	*J. E. Howard, Tests of Metals.* 1893, p. 283. From Poole...	36,820	.....	1.006	2.367	3.373	1.104	0.450	0.466	0.192	.....	0.010
34	*J. E. Howard, Tests of Metals.* 1894, p. 250. 12-in. shell....	34,450	.....	0.366	2.279	2.645	1.024	0.353	0.496	0.118	.....	nil
35	*J. E. Howard, Tests of Metals.* 1895, p. 181. 8-in. disappearing carriage.	39,540	.....	0.667	2.386	3.053	0.800	0.441	0.538	0.090	.....	.....
36	*J. E. Howard, Tests of Metals.* 1897, p. 282. "London, 1825."	35,420	.....	0.889	2.435	3.324	1.000	0.450	0.383	0.139	.....	.....
37	*J. E. Howard, Tests of Metals.* 1897, p. 282. Muirkirk pig, No. 4, high.	43,600	.....	0.573	2.358	2.931	0.700	0.654	0.340	0.085	.....	.....
38	*I. Ulke.* Cupola charcoal gun-iron.....	37,000	0.5–0.6	1.0–1.2	1.4–1.6	2.4–2.6	0.8–1.2	0.6	0.3	0.05	.....	.....
	Limits, lb. per sq. in	34,010–49,900	.....	0.366	1.4–	2.39–	0.8–	0.32–	0.057–	0.03–	.....	Nil to
	Limits, kg. per sq. mm	24.0–35.1	.....	1.276	2.96	3.93	2.99	1.512	0.773	0.192		0.160

1 to 3. *Year Book, Amer. Soc. Testing Materials*, 1915, p. 312. 4 to 6. *The Foundry*, 1907, vol. 31, p. 160. 7. *Iron Age*, 1908, vol. 81, p. 1926. 8. Specifications for breech loading mortars. *Rept. Chief Ordinance U. S. Army*, 1900, p. 400. The average tensile strength of the specimens from each end of each mortar must be at least 30,000 lb. per sq. in. (21.1 kg. per sq. mm.). The strength of one specimen from each end may be as low as 28,000 lb. per sq. in. (19.7 kg. per sq. mm.) but no specimen shall have a tensile strength greater than 37,000 lb. per sq. in. (26 kg. per sq. mm.). See also Kent, *Mechanical Engineers' Pocket-book*, Wiley, New York, 1899, p. 374. 9 to 14. *Beitrag zur Untersuchung des Gusseisens*, Stahleisen, Dusseldorf, 1913, p. 194. (All values given are high.) 15. G. R. Johnson, *Journ. Iron and Steel Inst.*, 1898, vol. 54, No. II, p. 203. 16. Private communication, April 15, 1895. 17 and 18. *Iron Age*, 1911, vol. 87, p. 1399. 19 and 20. *Mechanical Engineers' Pocket-book*, Wiley, New York, 1899, p. 374. 21 and 22. Private communication, April 1, 1902. 23 and 24. *Journ. Iron and Steel Inst.*, 1898, vol. 54, No. II, pp. 210 and 211. 25. Keep, *Cast Iron*, Wiley, New York, 1902, p. 136. 26. *Trans. Amer. Inst., Mining Eng.*,1888–9, vol. 17, pp. 472 and 473. Mr. Gridley afterward described two tests which gave 44,500 and 47,500 lb. per sq. in. in tensile strength respectively. 27 to 29. Private communication, April 30, 1902. In Nos. 27 and 28 the combined carbon and graphite are calculated by a formula. 30 and 31. Private communica-

TABLE 13.—*Concluded*

No.	Source and description	Mechanical properties: Tensile strength, lb. per sq. in.	Yield point, lb. per sq. in.	Elongation: per cent.	Elongation: in inches	Contraction of area, per cent.	Chemical composition, per cent.: Carbon: Combined	Carbon: Graphite	Carbon: Total	Si	Mn	P	S
	Malleable castings. Standard properties												
39	*American Society for Testing Materials*, Specification, 1915.	Not less than 40,000	not less than	2.5	2							not over 0.225	over 0.06
40	*E. Schoemann.* Made in open hearth furnace	44,230–48,640		3.9–4.5			trace		1.53	1.06	0.28	0.13	0.13
41	*W. P. Putnam*	39,638–50,849	28,326–41,792	7.03–10.15	2	10.51–20.92	to 0.14	0.40–1.94	0.40–1.94	0.74	0.24	0.148	0.041
42	*Kent, Master Carbuilder's Association*, 1891. Specimens 1.52 × 0.25 in.	34,700	21,100	2.0	4								
43	Specimens 2.0 × 0.78 in.	25,100	15,400	1.5	4								
44	Specimens 1.54 × 0.88 in.	33,600	19,300	1.5	4								
45	Specimens 1.52 × 1.54 in.	28,200		1.5	4								
46	*Kent*	32,000		2.0	4								
46A	*Toucsda.* Average strength of commercial	41,000											
	Limits, lb. per sq. in.	25,100–50,849	15,400–41,792	2.5–10.15	2								
	Limits, kg. per sq. mm.	17.7–35.8	10.8–29.4										
	Malleable castings. Unusually good properties												
47	*C. H. Gale*	60,000–70,000											
48	*H. R. Stanford.* Average of 3 tests, ½ in. diam.	55,100		5.2		5.5	0.15	2.80	2.95	0.69	0.19	0.140	0.064
49	" " " Average of 3 tests, ½ in. diam.	64,500		2.8		1.3	0.72	2.05	2.77	0.72	0.18	0.162	0.071
50	" " " Average of 2 tests, ½ in. diam.	69,100		4.0		2.6	0.31	2.35	2.66	0.76	0.19	0.145	0.075
51	" " " Average of 2 tests, ½ in. diam.	56,700		8.2		8.4	0.28	1.75	2.03	0.71	0.18	0.150	0.037
52	" " " Average of 3 tests, ½ in. diam.	51,600		7.0		7.7	0.32	2.54	2.86	0.75	0.19	0.129	0.023
53	" " " Average of 42 tests, ½ in. diam.	49,810		6.61		6.23	0.31	2.35	2.66	0.72	0.21	0.153	0.050
54	*S. B. Chadsey*	45,810–55,230		6.25–10.55	4					0.97–0.92	0.21–0.26	0.144–0.147	0.051–0.056
	Limits, lb. per sq. in.	45,810–70,000				1.3–8.4	0.15–0.72	1.75–2.80	2.03–2.95	0.69–0.97	0.18–0.26	0.129–0.162	0.023–0.075
	Limits, kg. per sq. in.	32.2–49.2											
	Steel castings. Standard properties												
55	*American Society for Testing Materials*, Specification, 1914. Hard, minimum	80,000	36,000	15.0	2	20.0							
56	Medium, minimum	70,000	31,500	18.0	2	25.0						not over	
57	Soft, minimum	60,000	27,000	22.0	2	30.0						0.05	0.05
58	*Kent*	60,000		10.0	4								
59	*British Standard* Specification, 1909. Minimum. Castings with wearing surfaces.	78,400		10.0	2								
60	Other general castings, and wheel centers	58,240		15.0	2								
61	*I. G. Johnson & Co.*, 1906 circular	60,000	32,000	34.0		50.0							
62	" " " " " , 1906 circular	65,000	34,000	33.0		50.0							
63	" " " " " , 1906 circular	70,000	35,000	30.0		45.0							
64	" " " " " , 1906 circular	75,000	40,000	27.0		38.0							
65	" " " " " , 1906 circular	80,000	43,000	24.0		36.0							
66	" " " " " , 1906 circular	85,000	45,000	20.0		32.0							
67	" " " " " , 1906 circular	90,000	48,000	18.0		25.0							
	Limits, lb. per sq. in.	30,000–90,000	27,000–48,000	10.0–33.0		20.0–50.0							
	Limits, kg. per sq. mm.	42.2–63.3	19.0–33.8										
	Steel castings. Unusually good properties												
68	*Tschernoff.* Annealed and air-cooled	86,690											
69	*Osann.* Remscheid (German) hard	135,100–149,300		1–6									
70	*Anonymous.* American, 1913	94,950	60,000	24.91		47.30			0.26				
71	" American, 1913	89,480	57,000	26.09		33.17			0.26				
72	" American, 1913	77,200	52,390	27.55		36.78			0.26				
73	" American, 1913	98,500	57,200	25.73		47.50			0.22				
74	" American, 1913	89,000	52,990	25.73		56.80			0.26				
75	" American, 1913	84,000	54,450	22.59		35.60			0.26				
	Limits, lb. per sq. in.	77,200–149,300	52,390–60,000	1.0–27.55		33.17–56.80			0.22–0.26				
	Limits, kg. per sq. mm.	54.3–105.0	36.8–42.2										

39. *Amer. Soc. Testing Materials*, Year Book, 1915, p. 309. Standard specifications Amer. Soc. Testing Materials, 1915. 40. *Stahl und Eisen*, 1909, vol. 29, p. 598. 41. *The Foundry*, 1911, vol. 39, p. 66. The numbers given for silicon, manganese, phosphorus, and sulphur are those in the original iron which contained 2.70 per cent. combined carbon. 42 to 46. *Mechanical Engineers' Pocket-book*, Wiley, New York, 1899, p. 376. 46A. Average of commercial. 10 per cent. elongation in 8 in. not uncommon. *Iron Age*, 1914, vol. 94, p. 949. 47. *Iron Age*, 1908, vol. 81, p. 1315. Shock resistance small. 48 to 53. *Trans. Amer. Soc. Civil Engineers*, 1895, vol. 34; through Johnson, *The Materials of Construction*, Wiley, New York, 1897, p. 115. 54. *The Foundry*, 1911, vol. 38, p. 51. The numbers given for silicon, manganese, phosphorus, and sulphur are those of the original white irons, which contained between 2.36 and 2.54 per cent. of total carbon. 55 to 57. Year Book, *Amer. Soc. Testing Materials*, 1915, pp. 155 and 156. Standard specifications, Amer. Soc. Testing Materials, 1915. 58. *Mechanical Engineers' Pocket-book*, Wiley, New York, 1899, p. 376. 68. "Manufacture of Steel," etc., translated by W. Anderson; Clowes, London, 1880, p. 11. 69. *Stahl und Eisen*, 1904, vol. 24, p. 719. 70 to 75. Private communications, 1913.

and hence the graphite content, as small as is consistent with the other conditions.

The data assembled in lines 15 to 38 of Table 13, §117, support these inferences perhaps as far as can be expected in the absence of information as to the most important variable of all, the arrangement of the graphite masses. They represent twenty-four cast irons with a tensile strength of 34,000 lb. per square inch or more, taken without selection from my records and other sources consulted. Out of the twenty-one cases in which the combined carbon content is given, it lies in all but six between 0.55 and 1.006 per cent. The content of graphite is indeed greater than might have been ex-

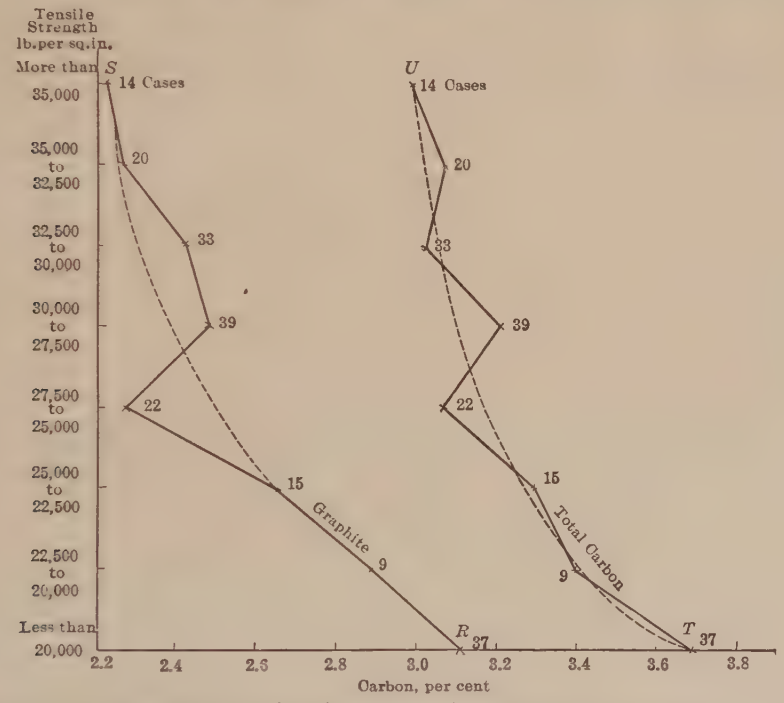


FIG. 19A.—The strength of cast iron increases as its composition approaches that of steel.

pected, lying between 2.31 and 2.84 per cent. in fifteen out of the twenty-two cases in which it is recorded, with one case above and six below these limits. The total carbon content in seventeen out of the twenty-two cases in which it is given lies between 2.74 and 3.32 per cent., with two cases above and three below these limits.

Summary of Table 13.—This may be summed up as follows:

Content of	Limits between which it lies in most of these cases	Number of cases			
		Total	No. within these limits	No. above these limits	No. below these limits
	Per cent.				
Combined carbon	0.55 and 1.006	21	15	3	3
Graphite	2.31 and 2.84	15	11	1	6
Total carbon	2.74 and 3.32	22	17	2	3

One suspects that this great content of total carbon represents process rather than product needs, and that the great graphite-content represents the necessarily large difference between this and the relatively small content of combined carbon 0.50 to 1 per cent., suited to give the product strength.

Further support is given by the curves *RS* and *TU* of Fig. 19A, made after the foregoing passage was written. They show that the strength of cast iron increases as it approaches high-carbon steel in composition, through a decrease in its total carbon content and graphite content. These curves *RS* and *TU* represent the tensile strength and composition of 180 cast irons taken without selection.[1]

These cases were divided into groups according to their strength, and then the average content of total carbon and of graphite of each group was plotted. The numeral written against each point shows the number of case which it represents.

As the tensile strength increases from	Less than 20,000 lb. per sq. in.	To more than 35,000 lb. per sq. in.	Or by more than 15,000 lb.
The total carbon content decreases from	3.69 per cent. to	2.98 per cent.	or by 0.71 per cent.
The graphite content decreases from	3.12 " " "	2.22 " "	" " 0.90 " "
And the combined carbon content increases from	0.57 " " "	0.76 " "	" " 0.19 " "

118. The Constitution of Cast Iron Illustrated Micrographically.—Black-heart malleable castings, because their content of combined carbon has been reduced by graphitization to a very small quantity, stand in Fig. 19 vertically above the low-carbon steels. Thus they are a conglomerate of (1) a metallic matrix which is the equivalent of such steel, and (2) the graphite resulting from the graphitization of the cementite which they contained when in the state of white iron, that is before they have been annealed. Fig. C of Plate 7, given me by Mr. E. Touceda, shows the structure of typical good black heart malleable castings. Note the white grains, which consist chiefly of ferrite, quite as a low-carbon steel does, and the rounded black areas, representing in turn a fine conglomerate of (1) graphite in impalpable powder filling the interstices within (2) a fine skeleton of ferrite. Figs. A and B of Plate 4, made by Prof. Wm. Campbell for this work, show the same thing on a larger magnification. Here the ferrite grains are recognized not only by their characteristic polygonal outline but also by their etch-figuring. That the dark area, S, in Fig. A represents the ferrite skeleton which traverses these graphitic areas will be explained in §123.

An ultra gray pig iron, with its carbon wholly graphitized, would have a structure like that of these malleable castings, with their compact graphitic areas replaced by these long sinuous flakes.

[1] Tests of Metals, made at the United States Arsenal, Watertown, from 1881 to 1911.

In gray cast-iron castings these relatively compact graphitic areas are replaced by long sinuous flakes of graphite, which evidently break up the continuity of the metallic matrix much more seriously. It is because of these flakes that gray cast-iron castings are so much weaker and more brittle than the steel to which their metallic matrix corresponds. Such graphite flakes are seen in Figs. D and E of Plate 4; C and E of Plate 5; F of Plate 6; E and F of Plate 12; and D and E of Plate 13.

Passing down our inclined column of Fig. 19, we have in the right-hand half of Fig. E, of Plate 4, an area the metallic matrix of which is equivalent to a hypo-eutectoid steel richer in carbon than those just considered, say a rail steel. Here the great white patches U,U′, are ferrite. They, with the great black graphite dyke which separates them, result from the graphitization of some mass of eutectic. This ferrite thus accompanies the zebra-striped pearlite at its left, and below it at the right of U′. It is this coexistence of ferrite with pearlite in this field that leads me to say that the matrix here is equivalent approximately to a rail steel, and hence that the alloy as a whole is equivalent to rail steel weakened and embrittled by coarse graphite flakes.

The part of Fig. C, Plate 4, which stands vertically above the letter R, in like manner shows in a malleable casting a region which corresponds to a hypo-eutectoid steel. In this region the black areas lack the star shape which they have near the right-hand edge of this same figure. This shape is one often taken by the graphite areas of malleable castings, as shown in Figs. A and B, Plate 4. In fact the rounded black areas above R represent pearlite, not resolved under this magnification. Hence this part represents a hypo-eutectoid steel, for it has the essential characteristics of such steel in being a conglomerate of pearlite with pro-eutectoid ferrite.

119. Examples of Variation in the Nature of the Matrix in a Single Specimen.—These two cases, Figs. C and E of Plate 4, illustrate a principle which is of the first importance here, that the matrix of any given cast-iron objects is likely to represent in different parts very different members of the steel-white-cast-iron series. Note that the left-hand part of Fig. C is of pure ferrite, and thus represents the very lowest carbon steels; the middle part contains rather more pearlite than ferrite, and thus represents hypo-eutectoid steel of about 0.60 of carbon, or rail steel; and the right-hand part again consists of pure ferrite plus graphite, and thus again represents the very lowest carbon steels, but now weakened by these graphitic areas.[1]

In Fig. E, Plate 4, the graphitization is far advanced in the right-hand part, which thus represents a hypo-eutectoid steel, whereas the left-hand looks as if graphitization had hardly started, so that this part is the equiva-

[1] This occurrence is common in black-heart malleable castings. Within the outer crust of oxide comes first a band of nearly pure ferrite, H in Fig. C of Plate 4, then next to the right a band of pearlite, R, then the regular "black heart," S, the typical mixture of ferrite with rounded or star-shaped graphitic masses enclosing ferrite skeletons.

lent of white cast iron, with its eutectic, T, and its pearlite, S, representing the primaustenoid.

So, too, in the upper and lower parts of Fig. E, Plate 5, the white metallic matrix is nearly pure ferrite, corresponding to an ultra low-carbon steel, whereas in the central part there is much gray unresolved pearlite between the white ferrite borders of the graphite flakes, so that here the matrix corresponds approximately to a rail steel.

120. Example of a Matrix Equivalent to a Hyper-eutectoid Steel.—Intermediate between the white cast-iron matrix at the left of Fig. E, Plate 4, and the hypo-eutectoid steel matrix at its right, is the hyper-eutectoid steel matrix of Fig. D. Here we recognize both pearlite and pro-eutectoid cementite. The lamellar pearlite which occupies the greater part of the field is recognized readily by means of its relatively narrow, slightly curved lamellæ of cementite, in a much more abundant ground mass of pearlitic ferrite. The pro-eutectoid cementite is in parallel needles, running as it were N 30° W, very like in tint to the pearlitic ferrite. In their parallelism over wide areas, their narrowness, and their straightness they correspond to the pro-eutectoid cementite in the hyper-eutectoid steels of Figs. G, H, and I of Plate 3; E of Plate 7; C to L of Plate 9; and B to F of Plate 11. Fig. D, Plate 4, has a few black graphite masses, or cavities left by their falling out, for instance the nearly horizontal streak in the lower part of the figure and the compact nearly vertical lens just at the right of letter L.

This coexistence of pearlite with pro-eutectoid cementite in the greater part of this field testifies that this metallic matrix is equivalent to a hyper-eutectoid steel, because such a steel consists essentially of this same complex. The presence of the graphite masses testifies that this alloy as a whole is a gray cast iron, which is thus a conglomerate consisting of two distinct parts, the metallic matrix equivalent to a hyper-eutectoid steel, and the graphite, which, as a foreign body, in effect breaks up the continuity of that matrix.

121. Inexactness of the Analogy between the Metallic Matrix and Members of the Steel-white-cast-iron Series.—These micrographs show us that the analogy between the matrix of gray cast iron and the members of the steel-white-cast-iron series is not strictly exact, because though their constituents ferrite, cementite, pearlite, eutectic, etc., are alike in structure and relative proportion, and also in chemical composition save for such minor differences as the greater silicon content of the ferrite and the greater manganese content of the cementite in cast iron than in steel, yet the reciprocal disposition or arrangement of these constituents in gray cast iron is not closely like that in steel. For instance, the needles of pro-eutectoid cementite in the gray cast iron, Fig. D of Plate 4, are thicker and more rounded than they usually are in a hyper-eutectoid steel of this carbon content. The ferrite and pearlite masses at the right of Fig. E, Plate 4, are shaped differently from those usually found in a rail steel. Such minor

structural differences are indeed to be expected, because of the difference in the conditions under which these constituents are generated.

One difference in these conditions is that the steel of most micrographs has been either forged or at least treated thermally in such a way as to give a new structure radically different from that which formed during the initial solidfication, whereas the cast irons have not. Hence what we see in the steels is a transformation structure, but in the cast irons a solidification structure. By giving the cast irons a suitable thermal treatment, they too can be given a transformation structure much closer to that of steel.

For instance, the cast iron of Fig. B, Plate 5 was heated to 900° for 15 minutes and cooled slowly in the furnace through the transformation range to about 600°, and then air cooled. This treament has superposed a transformation structure on the remains of the solidification structure. The great eutectic masses, T, T, T, are readily recognized, both by their spandril outline, concave to the bold convex dark austenoid, and by their honeycomb structure. The large dark austenoid mass which occupies most of the field is split up into three large regions separated from each other by the white islands of ferrite, U, U′. They thus form a rough network system, the ferrite islands representing the familiar ferrite network or shells of hypo-eutectoid steel, as in Figs. B and D of Plate 1 and the dark meshes representing the familiar pearlite kernels within those shells. The lower right-hand kernel is indeed tolerably lamellar pearlite, certainly lamellar enough to leave no doubt of its identity.

Internal evidence strengthens the inference that the pearlite and ferrite masses correspond to the ferrite shells and pearlite kernels of hypo-eutectoid steel. In steel we take each such kernel with its shell to represent one austenite grain, which in transforming has ejected to its outer surface the ferrite to which it has given birth in cooling through the transformation range. We have seen in Fig. A of Plate 1 how the orientation of the pearlite changes, from what was one austenite grain to what was another. A like change can be traced in Fig. B, Plate 5. Thus the structure of the pearlite in the lower right-hand grain, at the right of the ferrite island U′, differs strikingly from that in the two other main divisions of the pearlite.

122. The Foundry Products as Additional Examples of the Constitution of Gray Cast Iron.—Hydraulic cylinders. Because they stand, in virtue of their combined-carbon content of 0.80 to 1.20 per cent., vertically above the piano-wire and even higher carbon steels in Fig. 19, the metallic matrix which they contain is roughly equivalent to these hyper-eutectoid steels, broken up of course by the skeleton of graphite. Why are they given this constitution? Their matrix is made equivalent to a very high carbon steel because of the great strength which such steel has. Their graphite content is purposely kept low in order to lessen both its weakening effects and also the chance of the high-pressure water leaking through their walls, along

the surface of the graphite flakes, on the principle that the smaller the quantity of the graphite the less the likelihood of there being continuities of such flakes through the cylinder walls, giving the water a path of low resistance.

Machinery castings are usually much thinner than these massive hydraulic cylinders, and thus more exposed to danger of being broken by shock, or by being bent beyond their powers. Hence their matrix represents a steel poorer in carbon and hence more enduring of shock than the matrix of the cylinders. A further reason for giving them a lower carbon matrix is to make them softer, and thus to lessen the cost of machining them to their final shape, a cost which may form a large fraction of the total cost of the casting.

Radiator, stove plate, thin machinery castings, and black-heart malleable castings stand still further to the left, vertically over the rivet and boiler plate steels, to which they correspond in their content of combined carbon; and, with the exception of malleable castings, still higher vertically because of the greater degree to which their carbon has been graphitized in order to reduce their content of combined carbon to this small quantity, and thus to give them the greatest resistance to shock and the greatest ease of machining. Thus they consist of a metallic matrix equivalent approximately to a rivet steel, with its continuity broken up by the presence of much graphite. Cast-iron pipes are made of No. 3 iron. Its relatively large cementite content is relatively harmless here because no machining is done and the small size of its graphite flakes makes for imperviousness.

This great graphite content is given, not for itself—indeed it is a cause of weakness—but in order that, in spite of the low content of combined carbon, the total carbon content may be high, and that the metal shall thus be fusible and hence shall remain fluid, in spite of being cooled rapidly as it runs, by the chilling action of the walls of the narrow moulds in which these thin castings must needs be cast. It is given in short, not for product but for process purposes.

The lower position of the malleable castings in Fig. 19 results from the fact that, though their carbon is nearly completely graphitized, yet the graphite content is purposely made relatively small, so as to lessen its weakening and embrittling effect. This combination of a small content of combined carbon and a relatively small content of graphite of course implies a relatively small total carbon content, with a corresponding loss of fusibility, and hence increase in the foundry costs.

123. The Apparent Excess of Graphite in Malleable Castings.—The proportion of graphite represented by the rounded or star-like areas, for instance Fig. D of Plate 7, at first sight seems excessive, for we naturally take each of them to be made up of graphite alone, or to represent a hole left by the falling out of graphite. Taking into account the fact that there is probably about 2.30 per cent. of graphite, and that the density of iron (7.88) is about 3.5 times that of graphite (2.255), the area of graphite masses should

be about $2.3 \times \frac{7.88}{2.255} = 8$ per cent.[1] of the whole, whereas in fact it looks more like 40 per cent. of the whole. The explanation of this probably is that the graphitization has formed an extremely fine skeleton of ferrite, in and about which the fine particles of graphite lie; and that this combination is so much softer than the ungraphitized area of this same field that, in polishing the specimen, it has been rubbed away. Hence the section of the ferrite skeleton fails to be seen, because out of focus, or because smeared over with graphite, or for both reasons.

In support of this explanation I offer Prof. W. Campbell's micrographs, Figs. A, B and C, Plate 4, of graphite-bearing regions in malleable castings. The reason why the graphitic area S in Fig. A is lighter than the larger one at the bottom of the micrograph is that it has been brought more sharply into focus. That this area S, which but for this sharp focusing would have looked like a cavity, is neither a cavity nor a mass of graphite, is shown by the presence of this light metallic mass. That this mass represents a ferrite skeleton is further made probable by the fact that only by means of gentle polishing can it be detected. We infer hence that it is extremely weak, breaking down under hard polishing, as such a skeleton naturally would.

Metallic skeletons are seen in some of the dark areas in Fig. B of Plate 4. It is of special interest that certain parts of this skeleton, in particular the narrow light lines running along the axes of the dark spaces above grains T, U and V, are evidently continuations of the light boundaries which surround the remaining sides of those grains.

Further support of this explanation is given by the interpretation in §125 of Figs. C, D, and E of Plate 14, and in §304 of the rather similar black spots in the mottled part of a chilled casting, Fig. C of Plate 13, which on larger magnification, Fig. D, are seen to be crossed by relatively thin pearlite bodies. Though these look rather thick on this great magnification, yet as regards their resistance to the polishing, they are hardly more than webs. The pearlite body marked V, for instance, is less than three ten-thousandths of an inch (0.000,3 in.) thick. The more thorough graphitization in the

[1] Eight and not, as one at first believes, $(\sqrt[3]{8})^2$. A 3-in. cube, made up of twenty-six white 1-in. cubes set symmetrically about a central 1-in. red cube, is set on one of its faces, and three horizontal sections are taken, one through each of its three horizontal layers. These sections collectively disclose twenty-six white cubes and one red one, so that the proportion of red area disclosed is the same as the proportion of red to white by volume. Take the three sections vertically and the same result follows, and so it does with any number or direction of sections, if taken systematically. Pile other like 3-in. cubes symmetrically about the first and the same result follows, and so it does if we pile them unsymmetrically. In short, the proportion by sectional area equals the proportion by volume.

Sauveur came to this same conclusion in 1896, properly stipulating that exact identity would exist only under exact conditions of sampling. Such exactness of sampling is assumed in the foregoing. Sampling would not be mathematically exact except under the conditions which he stipulates in saying "Mathematically, however, the two ratios are equal only if the number of particles is infinite and the uniformity of distribution perfect." *Trans. Amer. Inst. Mining Eng.*, 1896, vol. 26, p. 879, footnote. See H. H. Campbell, "The Manufacture and Properties of Iron and Steel," New York, 2nd. edition, 1903, p. 406; 4th edition, 1907, p. 299.

manufacture of black heart malleable castings would naturally substitute ferrite for this pearlite. Indeed Mr. E. Touceda informs me that in these malleable castings, when they are imperfectly annealed, ferrite is replaced by pearlite.[1]

124. Other Cases in which the Graphite Masses are Compact. Johnson's Oxygenated Cast Iron.—A reminder of this compact grouping of the graphite in malleable castings referred to in §§118 and 123 is found in the compactness of the graphite masses in Mr. J. E. Johnson's oxygenated cast iron, Fig. F, Plate 4. Some of his cast irons similar to that here shown contain 0.01 to 0.07 per cent. of oxygen.

The cause of this compactness is considered in Chapter 11. This compactness is accompanied by a great increase in strength.

125. How Complete is the Discontinuity in Gray Cast Iron?—Such a micrograph as F of Plate 6 might easily give the impression that the flakes of graphite form a skeleton which comes very near to being fully continuous across the body of the casting. But were such discontinuity complete the strength of the whole would be only the strength of graphite, whereas in fact the strength of such a cast iron is often as much as one-fifth or even one-fourth or more of the strength of a steel of like combined carbon content. In short, cast irons seem to be stronger than the apparent extent of their discontinuity suggests.

Fig. F of Plate 12 explains this by showing that what seem to be graphite flakes completely isolating the metallic masses between which they lie, are in fact conglomerates, consisting of a ferrite skeleton with its interstices filled with graphite. Such a graphite skeleton binds together the metallic masses between which it lies. This skeleton is represented by the light metallic islets in the axial part of the great tapering shaft of graphite which runs nearly horizontally across the micrograph.

Other fields of this same specimen show this even more clearly.

The teaching of this micrograph is reinforced by that of Figs. C, D and E of Plate 14 which represents one field under different magnifications, and in the case of D and E under different focussings. The great graphitic band in Figs. C and E is seen clearly to be made up in very considerable part by metallic islets, indeed to be a sort of milky way.

The large white ferrite island T is seen in Fig. C to extend nearly across the graphitic band, the structure of the part within the band suggesting pearlite. So, too, the lamellæ of the pearlite on the side of the graphitic band opposite T seem in Fig. D to extend past the borders of the graphitic area.

The question "what has become of the three molecules of iron set free by the graphitization of the cementite, in those parts of the mass where the graphite is not bordered by ferrite," is answered here very clearly. This free iron exists as ferrite, or perhaps in part as pearlite, within the graphitic area, which is thus in fact a band not of graphite but of the conglomerate of ferrite + graphite resulting from graphitization of cementite.

[1] Private communication, July 3, 1914.

Two facts in connection with these Figs. C, D, and E show that the reason why the graphitic areas in cast iron usually look black instead of thus mottled is that the weak ferrite + graphite conglomerate is rubbed away in polishing the specimen so that it is out of focus. First, the focussing in Fig. D on the metal outside the graphite band fails to define the light areas in the band, whereas focussing on the band itself, as in Fig. E, so as to define these light areas, leaves the pearlite in the metal outside the band out of focus, though this very pearlite is seen in Fig. D to be sharply lamellar. This shows that the plane of the graphitic band differs from that of the rest of the surface.

Second, the borders of the graphitic area are obstinately jet black, and cannot be made to look white by any focussing, which is exactly what ought to happen if the surface of the graphitic band were thus rubbed down below that of the adjoining metal, for in this case the edges of this depression would naturally be rounded, and hence could not by any focussing be made to reflect the normal rays of light back into the microscope, and hence must always look black.

Further evidence in the same direction is given by the variations in the width of the black band with the variations in magnification and focussing. Taking the part immediately opposite the letter T, the band is wider in E than in D of Plate 14, and in D than in C allowing for the change in nominal magnification, differences which tally with the belief that these black borders are purely optical effects. Looking back to Fig. F of Plate 12, we note a like contrast between the jet blackness of the borders of the graphitic area and the milky way structure of its interior parts.

These observations show that what we usually call "graphite flakes" in gray cast iron really are like the graphitic areas in mottled cast iron and in malleable castings in being a conglomerate, consisting of a metallic skeleton with its interstices filled with graphite, a skeleton which may well bridge what at first looks like a weak graphite dike, and thus bind together the metallic bodies on either side of the dike. They tally perfectly with the theory that graphitization occurs after solidification. Yet in other cases, as in those of the great graphite masses which occur in the salamanders from the hearth of the iron blast furnace, the graphite seems to be perfectly compact and free from any iron skeleton. These may well represent the slower graphitization during a very long stay at the high temperature, which would permit the ferrite resulting from the graphitization of the cementite to migrate through surface tension completely out of the graphitic mass. And for that matter these thick graphite salamander flakes may have formed directly from the molten, and not through the graphitization of cementite.

126. Correspondence between the Steels and the Cast-iron Foundry Products in their Basis of Classification.—Though, as we have seen, the various grades of pig iron differ from each other primarily in their graphite content, the cases which we have considered show that the really valuable qualities of cast iron foundry products depend upon their content of com-

bined carbon, so that in this respect their basis or classification is the same as that of the steels.

It is true that the iron founder speaks rather of the grayness or whiteness of his iron, that is of the graphite content, than of its combined-carbon content, because the graphite is more conspicuous than the cementite in which the combined carbon is contained. But the graphite content is of importance not for itself but partly because it is an inverse index of the combined carbon content of the metallic matrix, in the sense that, for given total carbon content, the more graphite there is the less is the content of combined carbon; and partly as a necessary consequence of aiming to have not only a soft, shock-resisting, cheaply machined metallic matrix in the casting itself, the product, but also for the foundry process a material initially cheap, cheaply melted, and readily cast even into intrinsically difficult shapes, qualities which can be had only by means of a great content of total carbon. In some cases, too, the generation of graphite is useful in the process because, occurring as it does during and immediately after solidification, it causes a sudden expansion, which both lessens the tendency of the castings to tear themselves in pieces during solidification, and causes them to reproduce more accurately the shape of the pattern.

But these are benefits not to the product but to the process. In the product itself graphite is usually only the less of two evils. Given a great carbon content for the needs of the process, that carbon not only weakens and embrittles the product less, but hardens it less and thus makes it cheaper to machine, if present as graphite than if present as cementite.

This the justification of my assertion early in this chapter that in the last analysis the source and measure of the most valuable qualities are the same for the cast-iron products as for steel, the content of combined carbon or cementite. But it is doubtful whether any large fraction of our founders yet understand this. The art is in darkness.

Thus the primary difference between the various steels, whether castings or rolled or forged objects, and also that between the various classes of gray iron castings, is in their content of combined carbon; that between the various pig irons is in their graphite content; that between gray and malleable castings is the greater compactness and smaller total quantity of the graphite masses in the latter; while steel castings differ from both gray cast iron and malleable castings essentially in lacking graphite altogether, thus gaining greatly in strength and ductility. But this gain has to be paid for heavily, because it implies a much smaller content of total carbon in the molten metal, and hence much less fusibility and a higher casting temperature; the lack during solidification of the expansion which accompanies graphitization; hence greater contraction in solidifying and cooling jointly; and hence in turn endless special precautions to prevent the defects which this contraction tends to cause.

127. Relation between Gray-iron Castings, Malleable Castings, and Steel Castings.—The carbon content of steel castings habitually lies between

0.10 and 0.70 per cent., and like that of other steel products it is nominally allcombined.

Thus as regards their combined carbon content the steel castings richest in carbon correspond to the hydraulic cylinders and motor cylinders among the gray cast-iron castings; and those poorest in carbon correspond to the stove gray-iron castings, and to black-heart malleable castings.

Malleable castings differ from the gray stove plate castings in having their graphite brought into relatively harmless compact masses instead of the long continuity-destroying flakes of the gray castings, and in containing less graphite.

Steel castings differ from both in lacking graphite and its weakening and embrittling effect.

Thus when the product needs somewhat more but not greatly more strength and shock endurance than can be had in gray cast iron, the founder goes to the additional cost of making malleable castings, by suppressing graphitization during solidification and thus avoiding the long graphite flakes which form during solidification because of the mobility at that high temperature, about 1,135°, and instead bringing graphitization about at a much lower temperature, about 730°, by the long and costly annealing of the malleablizing process. In this compromise process he retains much of the cheapness of the casting process by retaining a relatively great total carbon content in his materials, and the fusibility and fluidity which follow, and yet lessens the damage to the product caused by the large graphite content which the combination of great total carbon content with small combined carbon content implies, by giving his graphite a less harmful form.

But when he really needs great strength and ductility he has to avoid the presence of graphite by making the total carbon content of his molten metal before casting substantially identical with that which the product itself needs to have in the combined state as cementite, that is, from 0.10 to 0.70 per cent., against some 2.70 per cent. in that for black-heart malleable castings and some 2.40 to 3.70 per cent. in that for gray-iron castings.

128. Condensed Statement of the Relation of Graphitization to Properties.—The generation of graphite during solidification leads to a sudden expansion of the metal in setting, so that it fills the moulds sharply. Hence for ornamental castings graphitization should go far. On the other hand, the presence of this graphite skeleton offers to liquids a path of relatively easy passage, or in other words lessens the imperviousness of the casting, so that for hydraulic presses graphitization should be restrained, and for radiators the formation of long flakes should be avoided.

Fluidity is favored by the presence of phosphorus and lessened by that of sulphur.

Considering a series of cast irons otherwise constant but with the degree of graphitization varying from zero at one end to completion at the other, the wholly ungraphitized or ultra-white cast iron is very hard and brittle because of its large cementite content, while the fully graphitized iron, such

as black-heart malleable and stove plate castings, at the other end, is a conglomerate consisting of a skeleton of graphite imbedded in a ferrite matrix which is in effect an ultra low-carbon steel.

As, starting from this end, graphitization becomes less and less complete, the matrix passes progressively through the stages of low-carbon, medium-carbon, and high-carbon steel, then becomes equivalent to a low-carbon cast iron, then to a normal, and finally to a high-carbon one.

In this progress we expect the matrix to reach its greatest strength when it is equivalent to a high-carbon steel, say when the cast iron contains 1 per cent. of combined carbon. But further reflection shows that the greatest strength is likely to be had with a somewhat smaller content of combined carbon, because with as much as 1 per cent. on an average of the whole, the local variations are likely to give certain areas so much more than 1 per cent. as to make them markedly eutectiferous and hence locally weak; and this local weakness is not helped materially by the strength of the matrix in the other parts, the strength of the whole being not so much greater than that of its weakest, which is its most eutectiferous part.

Graphite itself is a source of weakness and brittleness. Its presence is sought rather for process than for product purposes. For the combined carbon content desired, whether great for great strength as in the case of hydraulic cylinders, or small for softness and cheap machining as in the case of many small castings, is far less than the total carbon content needed to give fusibility and hence cheap founding. The difference must needs be represented by graphite, directly harmful to the product, but less harmful than the same quantity of carbon in the state of cementite.

When the needs of the product compel us to restrain the weakening effect of the graphite, and thus to restrain the graphite content, the rational way is to retain in the product its appropriate carbon content, and lessen the graphite content by lessening the total carbon content, even though this increases the foundry cost by lessening the fusibility, remembering that, for given content of combined carbon, the less the graphite the greater the strength. The relatively low total carbon content of hydraulic cylinder and gun irons illustrates this point. A further way of restraining the harm done by the graphite is to substitute malleable for gray castings, and according to Mr. Johnson to oxygenate the iron, either in his way or by the more costly step of using cold-blast charcoal iron, so as to make the graphite masses more compact.

I here condense some of these matters in Tables 14, 15 and 16.

TABLE 14.—CORRESPONDENCE BETWEEN THE DEGREE OF GRAPHITIZATION AND THE PROPERTIES OF CAST IRON

Matrix					Conglomerate or cast iron					
Name	C.C.	Hardness	Strength	Ductility	Gr.	Name	Hardness	Strength	Brittleness	Total C
Low C steel	0.06	Minimum	Moderate	Maximum	3.94	Ultra gray	Minimum	Moderate	Minimum	4
High C steel	1.0	Moderate	Maximum	Good	3.0	Open gray	Moderate	Maximum	Moderate	4
Ultra high C steel	2.0	Greater	Moderate	Small	2.0	Close gray	Greater	Less	Greater	4
White cast iron	4.0	Maximum	Minimum	None	0	White	Maximum	Minimum	Maximum	4

TABLE 15.—PROPER DISTRIBUTION OF CARBON BETWEEN THE STATES OF GRAPHITE AND CEMENTITE, TO GIVE VARIOUS PROPERTIES

Properties sought	Example of use	Distribution of carbon		Color of iron
		Cementite	Graphite	
Resistance to shock	Machinery	Little	.....	Open gray
Softness, easy machining	Machinery	Little	.....	Open gray
Sharpness of detail, expansion in solidifying.	Ornaments	.....	Much	Open gray
Fluidity (phosphorus)	Pipes	Moderate	.....	Close gray (No. 3)
Strength	Car-wheel centers	Moderate	.....	Close gray
Imperviousness	Hydraulic cylinders, radiators	.....	Very little	Very close gray
Hardness	Tread of car wheels	Much	.....	Nearly white

TABLE 16.—APPROXIMATE PROPORTION OF VARIOUS ELEMENTS IN CAST IRON FOR VARIOUS PURPOSES

Quantitative Control of the Distribution of Carbon between States of Graphite and Cementite, by Silicon, Sulphur, Manganese and Phosphorus	
	Silicon
Thick and hence slowly cooling machinery castings	1.50 to 2.25
Thin and hence fast cooling castings, ornamental castings	Up to 3 or even 3.4
Radiators, etc., needing density and low Gr	1 to 1.75
Car wheels, in order that C may shift easily from Gr to cementite	0.50 to 0.80
	Sulphur
Castings soft enough for machining	Not over 0.08
Very thin castings needing great fluidity	Not over 0.05
	Manganese
In good castings usually	0.40 to 0.70
Chilled car wheels	0.15 to 0.30
	Phosphorus
Machinery exposed to shock	Not over 0.50
Gas and water mains, and other castings not exposed to shock but needing fluidity.	Not over 1.60

If there is much sulphur or manganese, or if the castings are thin and hence cool rapidly, so that graphitization is restrained, the phosphorus content must be kept lower.

CHAPTER 7

SODIUM-NITRATE-WATER DIAGRAM

129. Introduction.—I urge the reader to master this chapter before proceeding further, unless indeed he is already familiar with the solidification of solutions. I urge this on him as a time and labor saving preparation for the study of what follows. Our whole conception of the constitution and properties of iron and steel, and more especially our understanding of heat treatment, rest upon the carbon-iron equilibrium diagram, or graphical representation of the course of solidification of the various carbon-iron alloys, and of the transformations which they undergo in their thermal treatment. With this diagram the reader ought to become so familiar that he not only reads and speaks but thinks in terms of it. Most of us have difficulty in getting clear conceptions of this diagram, essentially because the phenomena which underly it are unfamiliar. The present chapter aims to lessen this difficulty by familiarizing the reader with a diagram which, though it makes clear the principles of the carbon-iron diagram, yet is much easier to grasp, because it has to do with conditions which are much more familiar, the solidification of aqueous solutions. I lead through the simple to the complex.

This course has been followed by most writers. But whereas the aqueous solution which they have used for illustration is that of common salt in water, I use that of sodium nitrate or "Chili saltpeter" in water because of certain complications which, as we now know, arise in the freezing of common salt solutions, and prevent it from being closely parallel to the solidification of most molten alloys. But in the greater part of what follows "common salt" could be substituted for "nitrate" *mutatis mutandis*.

130. Summary.—The freezing of a saline solution, Fig. 20, occurs in general not at a single temperature, but through a range of temperature, and hence may have a marked mushy stage. The temperature at which freezing begins, AB, BC, varies with the degree of concentration of the solution. Starting with pure water and passing through a series of solutions progressively richer in the salt, the point at which freezing begins descends gradually from solution to solution, as from A to B. But there is a certain concentration called the eutectic concentration, B, with which freezing begins at a lower temperature than with any other concentration, and after this concentration has been reached any further enrichment leads to a progressively higher temperature of incipient freezing, BC.

Solutions richer than the eutectic are called hyper-eutectic, and those poorer, hypo-eutectic.

A solution with this eutectic concentration, or in short the liquid eutectic, differs from others in freezing completely like pure water at a fixed point instead of over a range of temperature.

Plotting, as in Fig. 20, concentration as abscissa and temperature as ordinate, the locus of the temperature and concentration of the mother-liquor, *AB*, *BC*, during freezing is called the liquidus, and that of the solid matter which freezes out, *aBc*, is called the solidus.[1] The temperature at which the freezing of any solution begins is that at which its ordinate intersects the liquidus; that at which its freezing ends is that at which its ordinate cuts the solidus. The temperature at which the freezing of solutions near in composition to the eutectic ends is usually identical with the freez-

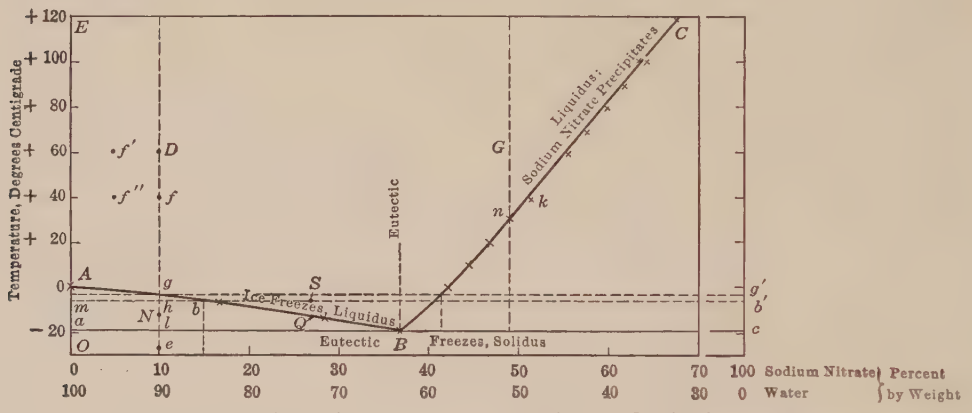


FIG. 20.—Freezing-point curve of sodium-nitrate (Chili-saltpetre)-water.

ing-point of the eutectic, and this identity may extend very far in either direction from the eutectic concentration, and indeed may, as in Fig. 20, include all concentrations from the weakest up to a point far beyond the eutectic.

A solution of eutectic concentration in freezing usually forms a composite or conglomerate mass, the solid eutectic, made up of separate particles of the constituents of the solution itself, as for instance particles of pure ice mixed with others of pure salt.

The freezing of all other solutions differs from that of the eutectic in being selective. That is, it consists in the progressive freezing out of that element, water in hypo-eutectic solutions and the salt in hyper-eutectic ones, which is present in excess of the eutectic ratio, thereby causing the residual mother-liquor to approach the eutectic ratio progressively, reaching it at the time when it reaches the freezing-point of the eutectic. Thus the freezing is of two parts, the pro-eutectic part which precedes the arrival of the mother-liquor at the eutectic concentration, and the eutectic part when the eutectic thus formed freezes. That which thus freezes out in the pro-

[1] In other cases, as in Fig. 23, §165, the horizontal part of the solidus extends only partway towards the axis of the diagram, here stopping at the point *E*. Thus whereas the solidus in Fig. 20 is the horizontal line *aBc*, in Fig. 23 it is the partly curved line *AEBC*.

eutectic period is called the pro-eutectic element, ice in the case of hypo-eutectic solutions, and the salt in the case of hyper-eutectic ones.

The composition and freezing-point of the eutectic are constant and independent of the initial concentration of the solution, and this composition is not usually in simple atomic proportion.

131. The Precipitation of a Salt from a Strong Solution.—If we hold water at + 40°C. in the presence of an excess of sodium nitrate, it dissolves that salt till it becomes saturated for this temperature, 40°, dissolving *k* or 51 per cent. If we now remove the water from the salt and heat it to say 100°, it thereby becomes undersaturated, because at this higher temperature it has the ability to dissolve more salt than the 51 per cent. which we have caused it to take up. If it is now cooled it remains undersaturated for this same reason, but the degree of undersaturation becomes progressively less, till when its temperature again reaches +40° the water has again become exactly saturated, in the sense that its capacity for dissolving salt is once more equal to the quantity of salt actually present in it. If we now start to cool it below 40°, some of the salt begins crystallizing out, because at temperatures below 40° the water cannot retain as much as at 40°, or in short because the solvent power of the water for nitrate, which means the solubility of the nitrate in water, decreases as the temperature falls.

Hence with further cooling there is further progressive crystallization of nitrate from the liquor, which at every degree that we pass impoverishes itself in nitrate by this out-crystallizing. That the water itself does not freeze when the temperature reaches 0°C. (32°F.) is familiar to us all, for who has not noted that even brackish water will not freeze on cold nights when pure water does?

But the freezing-point, though thus lowered by the presence of the dissolved salt, is not destroyed, and if the temperature is gradually lowered to −18.5°, the nitrate content of the solution simultaneously falling to 36.9 per cent. through continuing crystallizing, the whole freezes at constant temperature somewhat as pure water does, with this difference, that the solid which now forms is a conglomerate of minute particles of pure ice and other particles of pure nitrate. This conglomerate is called the *eutectic*, a substance of the greatest importance for the study of alloys.

132. The Liquidus.—If we plot, as in Fig. 20, the percentages of nitrate dissolved in the water, *i.e.*, its salt content as abscissa, and temperatures as ordinates, we get the curve *CB*, which represents the solubility of the salt in water for varying temperature, or the equilibrium line between solid nitrate and solution, called also the *liquidus*. Because *BC* thus represents the solubility of the salt in water, during the cooling of the solution the crystallizing out of the salt proceeds at such a rate that at each degree C. the quantity of nitrate remaining dissolved in the solution is represented by the abscissa of *BC* for that degree.

The precipitation of the excess of the salt in this case is exactly that

which occurs when any common hot saturated solution, for instance that of cupric sulphate, is cooled gradually. And here we may note that the freezing and crystallizing of ice, the precipitation or crystallizing of a salt from a solution which through cooling becomes supersaturated, and the freezing of a metal, are really one and the same thing, the passage from the fluid to the solid state. Water in freezing looks different from a salt crystallizing out, and both this freezing and this crystallization occur at temperatures so very much lower than those at which metals commonly solidify—the circumstances are so different—that we do not usually recognize that they are really one and the same thing. But this conception of the essential identity of the solidification of a metal, the freezing of water, and the precipitation of a salt, is very important for the student of alloys.

One teaching of the line *CB* is that, just as the quantity of the salt which the water will take up increases with the temperature, so the temperature at which the salt begins precipitating out rises and falls with the salt content of the solution.

133. The Freezing of a Weak Solution.—Up to this point we have been considering cases in which we saturated the solution with nitrate by holding it at a given temperature in contact with an excess of the salt. In this way we can readily cause the water to take up 60 per cent. or more of its weight of the salt. But let us now consider a different class of cases, in which we dissolve in our water at room temperature only a moderate quantity of the salt, say 10 per cent. of its weight, or far less than it is capable of dissolving. The solution is now greatly undersaturated.

No crystallizing out of the salt occurs as we heat such a solution and again cool it to the room temperature. Moreover, like sea water, it does not freeze at 0°C. but at a somewhat lower temperature, quite as the harbor remains open after the river which empties into it has frozen over.

We may follow the behavior of the solution during cooling by means of the ordinate *h* of Fig. 20. Because this is everywhere distant from the axis *OAE* by 10 per cent., the assumed nitrate content of our weak solution, it evidently represents this 10 per cent. solution at all different temperatures taken collectively, and any given point on it represents that solution at the temperature indicated on *OAE* horizontally opposite that point. Hence to follow a succession of such points down the ordinate is to follow the behavior of that solution in cooling.

The amount by which the presence of this 10 per cent. of nitrate lowers the freezing-point of the solution is in a sense 4°, and hence the beginning of freezing occurs not at 0° but at − 4°, point *g* on the curve *AB*. And now begins an extremely interesting process, which is strictly parallel to the crystallizing out of the salt from our concentrated solution studied in the preceding example. Just as the 51 per cent. solution began crystallizing out the salt when, in cooling, its ordinate fell to the line *CB*, so our present weak solution, when its ordinate in cooling reaches *g* on the curve *AB*, begins crystallizing out water in the shape of flakes of pure ice. More-

over, just as the crystallizing out of the salt from the strong solution proceeded at such a rate that the remaining solution had, at each temperature, the salt-content represented by the corresponding point on CB, so the crystallizing out of ice from the weak solution proceeds at such a rate that the water content of the residual solution is represented by the intercept between the curve AB and the 100 per cent. nitrate ordinate, the progressive freezing out of the flakelets of ice impoverishing the residual solution in water at such a rate that its water content is always represented by this intercept, and thereby enriching it in nitrate at such a rate that its salt content is represented by the abscissa of AB.

Thus whereas on reaching g, and just before any freezing begins, the salt content of the solution is 10 per cent., or g, and the water content is gg' or 90 per cent., when the temperature has sunk to b, enough ice has frozen out to reduce the water content to bb' or 85 per cent., and thereby to enrich the solution in nitrate to b or 15 per cent. Hence just as BC represents the salt content of concentrated solutions for varying temperature, so AB represents that of weak solutions.

When the temperature has fallen to h the mass as a whole consists of two parts, (1) the mother-liquor and (2) the flakelets of ice thus far frozen, and these two parts are represented both as to temperature and composition by the points b and m respectively.

This freezing out of ice continues till the temperature sinks to l or $-18.5°$ by which time the mother-liquor has been enriched up to composition B, 36.9 per cent. of sodium nitrate. But this temperature and this composition are those which the strong solution of the preceding example reached when it ceased crystallizing out the salt, and instead solidified at constant temperature as the eutectic mixture of particles of pure ice and others of pure nitrate. Hence our weak solution, having thus brought itself to the very conditions which led the strong solution to freeze as the eutectic, here behaves in exactly the same manner and freezes as the eutectic.

134. Mushiness in the Freezing Range.—What has just been said implies that the solution must pass through a mushy stage in freezing, like the mushy stage of salt ice familiar to those brought up on the seaboard. Thus at all temperatures between g or $-4°$ when the solution reaches the liquidus in cooling and freezing begins, and l or $-18.5°$ when it reaches the solidus and freezing completes itself, certain of its particles are of solid ice and certain others are still of liquid solution, and the whole consists thus of a mechanical mixture of solid and liquid particles, unless through difference of density the solid particles are able to separate themselves from the still liquid ones and float on their surface. In fact the water as it freezes habitually forms dendritic crystals which entrap the still liquid part, and thus mushiness does arise. Of course at the very beginning of the freezing the quantity of ice which has formed is too small, and the degree of mushiness is too small, to be recognized readily. And toward the end of the freezing the quantity of ice may be so great, and the consequent rigidity so great,

that we do not recognize readily the dual nature of the mass. But in some range of temperature between, and of greater or less width according to the conditions, this truly mushy stage can usually be recognized.

135. The Course of Solidification shown Quantitatively.—This course of events is shown in another way in Fig. 21. Here the lettering and the temperature or ordinate scale correspond to those in Fig. 20, but the abscissæ represent quantities of liquid and solid respectively. Till the temperature falls to *g* the whole is liquid. As it passes below *g* ice begins forming, and increases in quantity as sketched by the curve *gl*, till, in cooling past *l*, the solidification completes itself at constant temperature. The quantity of ice which froze out during the cooling from *g* to −18.5° is measured by *tl*, and that which froze at −18.5° by *lu*. This latter quantity is "eutectic," and *tl* is "pro-eutectic" ice, *i.e.*, that which came into existence before the eutectic.

In like manner the progress of freezing of a strong or hyper-eutectic nitrate solution is shown at *B* in Fig. 21. Starting with a solution containing

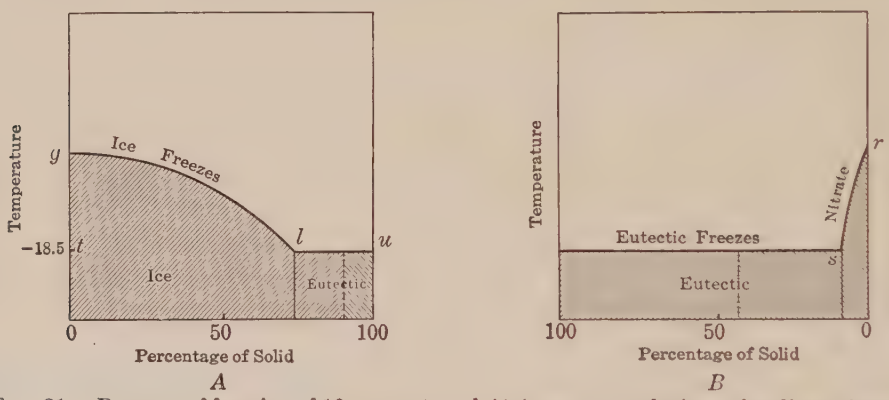


Fig. 21.—Progress of freezing of 10 per cent. and 41.4 per cent. solutions of sodium nitrate in water.

41 per cent. of salt, and therefore starting to crystallize out salt at −4°, which is the temperature at which the hypo-eutectic solution of *A* started to freeze, the progress of the crystallizing out of the salt is shown by *rs*. Its abscissa at each temperature measures the quantity of the salt which has crystallized out on falling to that temperature. As in Fig. *A*, the eutectic freezing occurs at −18.5°; but there is this striking difference, that because of the steepness of *BC*, the quantity of the salt crystallized out in cooling through this 14.5° range from *r* to *s*, is only about $\frac{1}{10}$th that of the ice which froze out from the weak solution in cooling through this same range.

On reheating the completely frozen weak solution this course of events is passed through in the reverse order. The eutectic re-melts at temperature *l*, and the mother-liquor thus formed gradually licks up the still unmelted ice with further rise of temperature, at such a rate that at each temperature its composition is represented by the point on *AB*, Fig. 20, cut by the abscissa of that temperature, and the quantity of ice by the abscissa of *gl* of Fig.

21, decreasing suddenly from 100 per cent. to *tl* at −18.5°, and thence decreasing according to the abscissa of *lg*, till on reaching temperature *g* the re-melting is complete, and the cycle rounded out. We may now generalize from these two cases.

136. Recapitulation. Meaning of the Sodium Nitrate-water Diagram, Fig. 20.—Any given solution of a salt in water, instead of having a definite freezing point as pure water and other pure substances have, has a freezing range; in other words its freezing covers a considerable range of temperature.[1] This is true of solutions in general, including not only the common aqueous solutions but also molten steel and cast iron, which may be regarded as a solution of carbon or of an iron carbide in molten metallic iron. Beyond this, the more salt the water contains, up to the eutectic ratio, the lower is the temperature to which it must be cooled in order to start freezing. But the temperature at which freezing completes itself is the same for all the different degrees of concentration of any given salt in water.

The freezing-point curve or equilibrium diagram of the sodium nitrate-water solutions in Fig. 20 assembles the data touching these matters. The lines *AB* and *BC* show the temperature at which freezing begins for any given percentage of the salt initially present, dissolved in the solution, and the line *aBc* shows that at which freezing ends.

137. Difference between the Freezing of a Pure Metal or Chemical Compound, and that of a Solution.—Let us now take up certain of these points in more detail.

When a vesselful of pure water is cooled to 0°C, its various concentric layers freeze at one and the same temperature, 0°, and their freezing is of course unselective, *i.e.*, the composition of the ice which forms is not only the same in all the successive concentric layers, but exactly the same as that of the water from which it freezes. So with other pure metals and pure chemical compounds. But solutions, in freezing, behave much less simply.

138. First, the Freezing of a Solution is a Differentiation.—Though water at the room temperature can dissolve sodium nitrate in any proportion up to about 49 per cent., and though many solids can, while remaining solid, contain other substances as truly dissolved in them as if they were liquid, solid gold containing solid silver truly dissolved in it, yet ice

[1] This assertion that a solution of a given percentage of a salt has no true freezing point may at the first glance offend our natural conceptions of the fixity of the laws of nature. Let me therefore hasten to add that, in a sense, it has such a fixed freezing-point, though in another sense it has not. A 10 per cent. salt solution has indeed a fixed point at which it begins to freeze, and it would all freeze at that one point, just as pure water all freezes at 0°, provided that it could be kept as a 10 per cent. solution. But, for reasons which we shall see in §139, the moment the solution begins to freeze, it inevitably begins to shift its degree of concentration, in other words to get a new and greater percentage of salt, and so to acquire a new freezing-point, which actually is at a lower temperature. Thus the fact that any given solution has a freezing range instead of a definite freezing-point reflects the fact that, as freezing progresses, the percentage of salt in the part remaining liquid constantly increases, so that the series of temperatures which make up its freezing range are in fact the freezing-points for the series of increasing percentages of salt which it has in succession, in the course of freezing. This will be explained at length in §140 and the following sections.

cannot retain any nitrate thus dissolved, nor can solid nitrate retain any water dissolved in it and yet remain solid. There may be mechanical entanglements and conglomerates of ice and nitrate of any degree of fineness and intimacy; but when you finally come to the true particles of ice themselves, they are saltless, and those of nitrate are anhydrous. Hence a completely frozen nitrate solution is a conglomerate of certain particles of pure saltless ice and other particles of pure anhydrous nitrate.[1] Thus freezing is a process of differentiation, splitting up an initially homogeneous solution into a conglomerate or mechanical mixture of its components.

139. Second, their Freezing is Selective.—This is another aspect of the same thing.

In calling freezing "selective" I mean that it proceeds by selecting from the liquid mother mass certain elements or compounds either alone, or in a proportion greater than that in which they existed in that mass before freezing began. In this present case freezing proceeds by selecting from the solution certain particles of pure water and freezing them out as pure ice. In actual practice the flakelets of this ice usually retain some of the solution mechanically entangled among them so that in this way the selection may appear to be incomplete, though the true solid particles themselves are of pure saltless ice.

140. Third, Depression of the Freezing-point.—Neither the solution as a whole nor the water in it freezes when the temperature falls to 0°. Moreover, every addition of the salt, up to 36.9 per cent., lowers the point at which the freezing of the solution begins, as is shown by the continuous slope of the curve AB in Fig. 20. This is but a single instance of the general law, that the dissolved substance or solute lowers the freezing-point of the solvent, in this case the water; or rather lowers the temperature at which freezing begins.

141. Fourth, their freezing covers a considerable range, as indeed follows from its being selective. For as the water which first freezes to flakes of pure ice, thereby rejects the salt which it had contained; and as this salt thereby concentrates progressively in the central region of still unfrozen water or "mother-liquor;" and finally as this consequently grows progressively richer in the salt, it thereby acquires a progressively lower and lower freezing-point; or in other words as the enrichment of the mother-liquor in nitrate goes on, a lower and lower temperature is needed to induce the further freezing out even of pure ice. Hence the considerable range of temperature covered by freezing.

142. Fifth, Ratio of the Two Parts.—At any given instant during freezing, as for instance when the temperature has fallen to point h, Fig. 20, the

[1] In certain cases the salt which freezes out is hydrated, but for simplicity we may pass these by. Further, we may as philosophers question whether any two things are absolutely insoluble in each other, and hold it more probable that all cases of apparently complete insolubility are in reality only cases of extremely small or negligible solubility. But this consideration we may omit here for simplicity.

mass consists of two parts theoretically distinct, the flakes of ice already frozen which, because they are of pure ice are represented by the point m; and second the still unfrozen solution, which is represented by the point b, because at every instant during freezing the solution is at its then freezing-point as represented by the curve AB. The proportion of ice to solution at any given temperature will be $hb:hm$, as is evident when we note that, at temperature g just before the first flakelet has frozen, that ratio is zero because there is no ice, and that when the whole has just frozen at temperature B the ratio is infinity, because there is no solution left; and further that at every intermediate instant the average composition of the whole must remain constant at 10 per cent., because all the initial nitrate and water still remain present; and in order that the average shall remain constant the ratio of the two parts must vary in this proportion. Or more directly, the percentage of ice is shown by the abscissæ of curve gl of Fig. 21.

143. Sixth, Freezing of the Eutectic Solution.—When a hypo-eutectic or weak solution has, through the freezing out of pure ice, enriched itself in nitrate to the eutectic ratio of 36.9 per cent., or when a hyper-eutectic or strong solution has, through the crystallizing out of the salt, impoverished itself to this same ratio, as each must needs do on reaching temperature B, the remaining solution freezes as a whole at constant temperature, $-18.5°C$, without further selection, yielding the eutectic mixture of alternate plates of ice and of nitrate.

Thus the freezing is of two parts, the selective freezing as the temperature sinks from g to l; and the unselective freezing of the eutectic at the constant temperature B.

During selective freezing, all the water in a hypo-eutectic solution, and all the salt in a hyper-eutectic one, which is present in excess of the eutectic ratio, freezes out at progressively falling temperature, so that this might be called the excess freezing, or better the pro-eutectic freezing period, as distinguished from the eutectic freezing period during which the residual mother-liquor freezes unselectively and at constant temperature as the eutectic.

144. Seventh, the Lower End of the Freezing Range is at Constant Temperature, Irrespective of the Initial Percentage of the Salt.—This follows from the fact that when the mother-liquor has been enriched up to the eutectic ratio of 36.9 per cent. of sodium nitrate, freezing ceases to be selective and the eutectic mother-liquor now freezes at its own freezing-point. This of course is independent of the initial percentage of the salt, because the eutectic is simply the solution of the lowest freezing-point among the whole series of possible sodium nitrate-water solutions, and the composition which has that lowest freezing-point is evidently independent of the initial percentage of the salt. Because its composition is thus independent, its freezing-point must be also.

145. Eighth, Double Meaning of the Liquidus or Line AB.—From the foregoing it is clear AB has a double meaning. For a series of solutions, it

is the temperature at which freezing begins; for a given solution it represents the percentage of nitrate which the mother-liquor contains at each successive temperature during selective freezing.

Starting with a solution at temperature 60°, and containing 49 per cent. of sodium nitrate, and therefore represented as to both temperature and composition by *G*, Fig. 20, and following its cooling, when the temperature reaches *n* freezing begins by the precipitation of particles of pure nitrate, and the consequent impoverishment in nitrate of the remaining mother-liquor. This shows us that the freezing out of ice from the 10 per cent. nitrate solution and the precipitation of the salt from the present one are exactly parallel, the differentiation of an initially homogeneous solution into two parts, a solid part and residual liquid part. We should not be confused by the fact that we habitually speak of the solidification of water as freezing, and the solidification of nitrate or other dissolved substance as precipitation, for as we have seen crystallization is to a salt exactly what freezing is to water.

As the temperature falls progressively, more and more nitrate precipitates out, and the correspondingly impoverished residual mother-liquor draws nearer and nearer to the eutectic percentage, 36.9, reaching that percentage when the temperature falls to −18.5°, *B* and thenceforth freezing unselectively as the eutectic.

A solution of exactly 36.9 per cent. of sodium nitrate does not freeze at all till its temperature has sunk to −18.5°, when it freezes as a whole, *i.e.*, unselectively as the eutectic.

146. To sum this up, the freezing of the solutions of sodium nitrate in water is selective; it brings the mother-liquor to the eutectic salt content of 36.9 per cent., by the progressive freezing out of ice from solutions with less than 36.9 per cent. of the salt, and of nitrate from those with more than 36.9 per cent. Selection ceases when the solution has reached 36.9 per cent., and the residual solution then freezes unselectively to form the solid eutectic. *The equilibrium or freezing-point curves which represent these phenomena consist of an underscored V*, the two branches of the *V* representing (1) the temperature at which the freezing of any given solution begins, and (2) the salt content of the residual solution at each temperature during that freezing, so that that solution as it were slides down *AB* or *CB* to *B*. The underscoring *aBc* represents the constant temperature at which freezing ends in all cases by the freezing of the eutectic.

147. Eutectic.—The fact that, no matter what the initial concentration of the solution is, the freezing of the eutectic occurs at one and the same temperature, and the mother-liquor which there freezes has the constant composition of 36.9 per cent. of sodium nitrate shows that this percentage composition has some important relation. This ratio of nitrate to water is the "eutectic" ratio, the ratio which gives the lowest freezing-point, *i.e.*, the greatest fusibility of the series of solutions, or, in the case of metals, the

greatest fusibility of the series of alloys of the two or more metals which we have in hand.

148. Properties of the Eutectic.—To recapitulate, the striking properties of the eutectic are:

(1) That its composition and hence its freezing-point are constant, no matter what the initial ratio between the solvent and the solute, *i.e.*, the composition of the solution or alloy; (2) that its composition is not, save occasionally and as it were by accident, in simple atomic proportions, because it is only the most fusible mixture and because there is no reason why the maximum fusibility should correspond to any definite atomic ratio; (3) that it consists of a conglomerate of distinct particles of the two components, mechanically mixed.

The solution or alloy which, as a whole, has this composition corresponding to the greater fusibility is the eutectic of the series, identical with the eutectic which is generated by the selective freezing of solutions or alloys of other compositions. Eutectic means literally "well-melting," from the Greek Ευτεκτος.

149. Reasons for the Properties of the Eutectic.—There are extremely simple reasons why the eutectic formed in the freezing of nitrate solutions should have the properties which have just been enumerated, surprising as those properties at first seem. These reasons are readily seen on following the freezing of a solution of say 15 per cent. of sodium nitrate in 85 of water. To this end let us again follow the progress of freezing. It begins with the freezing out of a little pure ice; this enriches the residual mother-liquor in the salt; and this in turn lowers the freezing-point of that mother-liquor; *i.e.*, the temperature to which we must cool it to induce further freezing of ice. This joint freezing out of ice and lowering of the freezing-point go on progressively through the ice-freezing period, a residue of mother-liquor persisting unfrozen, because as fast as the temperature falls, so fast does the freezing-point of the mother-liquor fall, thanks to its simultaneous enrichment in nitrate, so that at each instant the mother-liquor is at its freezing-point.

But there must be an end to this: there must be a limiting temperature at which no mixture of nitrate and water can normally remain liquid, and that limiting temperature is the freezing-point of the most fusible, *i.e.*, the most easily liquefiable mixture of nitrate and water, viz., that with 36.9 per cent. of nitrate, freezing at $-18.5°C$. On reaching this temperature and composition the mother-liquor itself should freeze. For, just as it was the progressive enrichment of the mother-liquor in the salt that caused it to remain unfrozen in spite of the progressive cooling, simply because that enrichment gave it a lower freezing-point; so when the enrichment in the salt reaches and tends to pass 36.9 per cent. it should fail to keep the mother-liquor from freezing, because further enrichment in the salt, instead of lowering, would raise the freezing-point; instead of keeping the mother-metal from freezing it would make it freeze all the sooner. At every instant in the cooling until this 36.9 per cent. is reached, the mother-liquor is at its then

freezing-point, and is kept from actually freezing by the simultaneous fall in that freezing-point caused by the freezing out of water and consequent enrichment of the mother-liquor in nitrate. The mother-liquor at each instant is on the point of freezing, but in starting to freeze renders itself unfreezable at the existing temperature by thus as it were ejecting from itself a little water and so making itself richer in nitrate. But when it has reached 36.9 per cent. of nitrate and simultaneously reached the freezing-point of that mixture, freezing cannot normally be deferred by further differentiation, because any change in composition would now raise the freezing-point; as the mother-liquor is now at its freezing-point, to raise its freezing-point without raising its temperature would freeze it. Hence, further selection offers no escape from freezing.

Hence we see why the composition and freezing-point of the eutectic are constant and independent of the initial proportions of the salt and water. For, no matter what those proportions, since the condition which enables the mother-liquor to remain unfrozen through progressive cooling is that the change in its composition (which, through the freezing out of one component, accompanies that cooling), leads to a lower freezing-point; so this permissive condition ceases when the 36.9 per cent. is reached simultaneously with the freezing-point for that limit, simply because no further change in composition can now yield a lower freezing-point, *i.e.*, no possible combination of nitrate and water can normally remain liquid at this temperature, a condition of affairs evidently wholly independent of the initial composition. Of course, the nearer the initial composition of the liquor is to this 36.9 ratio, the less will be the excess of water (or nitrate) to be frozen out before reaching this ratio, and the larger will be the quantity of the eutectic which will form; but the composition and freezing-point of the eutectic are clearly wholly independent of this.

150. Why the Eutectic is Composite.—That the eutectic formed on freezing sodium nitrate solutions consists of alternate masses of nitrate and of ice simply represents the assumed fact that solid ice and solid nitrate are insoluble in each other, soluble as nitrate is in liquid water. (For brevity I purposely assume here that this insolubility is complete.)

That though liquid water dissolves nitrate, nitrate and solid water, *i.e.*, ice, are insoluble in each other, necessarily implies that when the eutectic freezes they must (normally) cease to remain dissolved in each other, *i.e.*, they must mutually expel each other, so that each isolates itself from the other. Hence their existence as distinct particles. The constituents of certain alloys, for instance those of gold with silver, are indeed completely soluble in each other. Such a series might have a member more fusible than either pure metal, and this would be an uncomposite eutectic.

151. Why the Eutectic is not of Simple Atomic Proportions.—That the composition of the eutectic is not in simple atomic proportions is natural enough. Its composition is that of the solution (or molten alloy) of lowest freezing-point; and we see no reason why this lowest freezing-point should

correspond to any simple atomic ratio. It is the freezing-point of a homogeneous liquid solution which we are considering, rather than the melting-point of a solid, which we know to be a conglomerate mass. The very indefiniteness and indeterminateness of composition of solutions, and their freedom from marked critical points, seem opposed to the idea that their lowest freezing-point, or indeed their maxima and minima in general, should correspond to simple atomic proportions.

152. Arrival at Eutectic Composition and Freezing-point are Simultaneous.—That the arrival at the eutectic composition and freezing-point must be simultaneous is simply a single instance of the general condition which we have traced during freezing; for we have seen (§136) that at every instant during freezing the mother-liquor is beginning to freeze, *i.e.*, is at its then freezing-point. That its freezing is not completed at that temperature is simply because the selective change of composition which accompanies the progress of freezing is to and through a series of compositions each corresponding to a lower freezing-point than the preceding.

153. Eutectic Plus Excess or Pro-eutectic.—Whatever be the initial percentage of the salt in our sodium nitrate-water solution, when it is frozen it will contain a eutectic of 36.9 per cent. of salt. If there is initially present say 40 per cent. of the salt, or an excess of 3.1 per cent. over this eutectic ratio of 36.9 per cent., then the mass when frozen will be a conglomerate, containing in addition to the eutectic the salt which crystallized out in the early part of the freezing, while the solution was impoverishing itself in nitrate and thus approaching the eutectic ratio, in short the salt which crystallized out before the eutectic, or the *pro-eutectic* salt, in separate crystals, mechanically intermixed with the eutectic. If on the other hand there is say 15 per cent. of the salt initially present, so that there is 21.9 per cent. excess of water over the eutectic ratio, then the mass in freezing as a whole will form a conglomerate consisting of the eutectic as before, plus this 21.9 per cent. excess of mechanically intermixed crystals of ice, which froze before the eutectic, or in short the *pro-eutectic* ice. Thus, the frozen mass will in every case be a conglomerate, consisting of *a eutectic plus pro-eutectic substance*, which will be either nitrate or ice, as the case may be. In the liquid state, if water is in excess over the eutectic ratio we say that we have nitrate dissolved in water; if the salt is in excess we may say that we have water dissolved in the salt. The pro-eutectic substance of the solid state is the solvent in the liquid state, in this view. (See note to §160.)

154. Definitions. Hyper- and Hypo-eutectic.—The conglomerate or solid, or impure ice, formed by the freezing of a weak solution with less than 36.9 per cent. of sodium nitrate is hypo-eutectic, *i.e.*, it contains less nitrate than the eutectic percentage. Ice formed by the freezing of a strong solution with more than 36.9 per cent. of the salt is hyper-eutectic, *i.e.*, it contains more nitrate than the eutectic.

Hypo-eutectic differs from hyper-eutectic ice in being a mixture of the eutectic with an excess of pure ice instead of with an excess of nitrate.

155. Pro-eutectic Salt and Ice.—The ice which freezes out from a weak solution, and the salt which crystallizes out from a strong one, during the cooling down toward the eutectic freezing point of $-18.5°$ are called pro-eutectic ice and salt respectively, and the ice and salt together which compose the eutectic may be distinguished as eutectic ice and eutectic salt.

156. Liquidus and Solidus.—The lines AB and BC which represent the successive temperatures and salt contents of the mother-liquor remaining during the progress of freezing of a series of solutions are called collectively the liquidus. The lines Aa and aBc which represent the successive temperatures and salt contents of the part which has solidified are called the solidus. For instance, during the freezing of our 10 per cent. sodium nitrate solution, the fact that pure ice freezes out at the temperature sinks from $-4°$ to $-18.5°$, g to l, is represented by the part ma of the axis OAE, the salt content of the frozen part here being 0. The fact that this selective freezing is replaced by the freezing out of the eutectic while the temperature remains constant at $-18.5°$, and that this addition of the eutectic to the pure ice first formed progressively raises the salt content of the whole solid part, till, at the end of solidification, the solid as a whole has the 10 per cent. of nitrate with which we started—this fact is represented by the part al. The composition of the layers which are successively deposited, after descending thus along the axis from A to a, leaps across to B, when the unselective freezing of the eutectic replaces the selective freezing of pure ice.

To sum this up the *liquidus traces the history of the liquid or mother-liquor, the solidus the history of the precipitated or frozen or solidified part*, during freezing and melting, a conception which should be prominently in mind.

157. The Eutectic Line Runs Quite to the Axis.—This is clearly implied by the assumption that the ice which freezes out during the selective freezing is wholly saltless. For, no matter how minute the quantity of the salt initially present is, if absolutely none of this freezes out, the whole of that minute quantity must continue accumulating in the mother-liquor, continually enriching it in nitrate and at last bringing a minute quantity of it quite to the eutectic salt content of 36.9 per cent. and so generating a minute quantity of eutectic. But if that quantity is extremely minute, its presence may escape actual detection and identification.

158. Perfect and Imperfect Selection in Freezing.—This inference that the eutectic freezing line runs completely across the diagram is based on the assumption that the selection in freezing is perfect, *i.e.*, that the ice which freezes is perfectly pure and can contain no nitrate dissolved in it. But if ice were capable of retaining a small quantity of the salt dissolved in it and could yet remain solid ice, then the first particles to freeze out need not be pure ice but ice containing nitrate. In that case, if the solution initially contained only a small quantity of the salt, this might be so far retained by the ice as it freezes that the enrichment of the mother-liquor in the salt would never reach the eutectic ratio, and consequently that no eutectic would form. In that case the eutectic line would not run as far as the axis

OAE, but would stop at that salt content with which the enrichment barely brings the last drop of mother-liquor up to the eutectic content, as in Fig. 23.

159. Equilibrium Diagram.—These curves represent the ideal conditions when there is no lag, but equilibrium is fulfilled exactly. Again they represent equilibrium in this sense: that for given temperature *m*, *b* represents the sodium nitrate content of the liquid which can remain in equilibrium with ice represented by *m*, without either dissolving that ice or adding to it. In the language of the phase rule, for given temperature these curves represent the degree of concentration of the liquid and solid phase respectively which correspond to equilibrium.

160. The Liquidus is a Solubility Curve or Saturation-Point Curve.—The very fact that, in the cooling of a 49 per cent. nitrate solution, pure salt barely starts to freeze out when the temperature falls as far as *n*, shows that at *n* the solution is just saturated with salt, or in other words that, at temperature *n*, 49 per cent. of the salt just saturates water, and water is able to

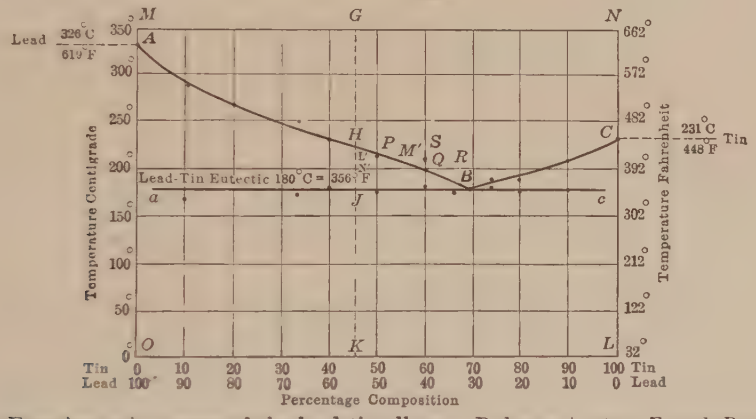


FIG. 22.—Freezing-point curve of the lead-tin alloys. Roberts-Austen, *Fourth Report Alloys Research Committee*, 1897, Fig. 5, Plate 4.

dissolve only just 49 per cent. of the salt; so that a series of such points form a saturation point curve or a solubility curve according to our point of view.[1]

[1] To the reader who replies that what is sauce for the goose is sauce for the gander, and that if *BC* is the solubility curve of the nitrate in water, then *AB* should be the solubility curve of water in nitrate, I can reply only that the terms "solvent" and "solute" seem unphilosophical. This is seen by taking an extreme case. Admitting that *AB* is thus correctly described, and considering a solution with 0.000,01 per cent. of the salt and 99.999,99 per cent. of water, we have the extraordinary consequence that 1 part of the salt is able to dissolve practically 10,000,000 parts of water. If the reader does not think this proportion ridiculous, he can reach any proportion which will satisfy his idea of the ridiculous by decreasing suitably the assumed percentage of the salt. Or, if he is confused by the idea of the water dissolving in the salt, let him turn to the perfectly parallel case of the lead-tin alloys, the freezing-point curve of which is shown in Fig. 22. When alloy *G* cools to *H* lead begins to freeze out. In this alloy there are 55 per cent. of lead and 45 of tin; the fact that on cooling to *H* lead begins falling out is proof that above *H* the 45 parts of tin dissolved the 55 of lead. Hence *AHB* is the solubility curve of lead in tin. But in an alloy of 99.999,99 parts of lead and 0.000,01 parts of tin, we have as before at temperature *A* 10,000,000 parts of lead dissolved in 1 of tin.

Such cases may be met by regarding the rôles of solvent and solute as not only reciprocal but identical, and the distinction between them as useful only for purposes of description.

161. A Dissimilarity Between Hypo- and Hyper-eutectic Freezing.— During the pro-eutectoid freezing period, that which solidifies from a hypo-eutectic solution is ice, floating at the surface, coherent, bulky, spongy, and locking up in its pores much of the still truly liquid mother-liquor. From its very incipiency freezing declares itself prominently as freezing.

How strikingly this process differs in appearance from the corresponding pro-eutectic freezing of a hyper-eutectic solution, *e.g.*, of common salt. Here that which solidifies is salt, small in quantity, compact, incoherent, and falling like a powder to the bottom of the vessel. The obscurity under which freezing here hides itself equals the clearness with which the freezing of a hypo-eutectic solution asserts itself. This contrast is reflected in the diagrams *A* and *B* of Fig. 21, by the much greater weight of ice formed in *A* than of the salt formed in *B* during the pro-eutectic freezing. But the contrast to the eye is even greater than in these diagrams, because the ice is so much lighter than the salt.

The eye recognizes that freezing begins with the beginning of the pro-eutectic freezing of the hypo-eutectic solution, but in the freezing of a hyper-eutectic brine the eye does not recognize that freezing has begun till the eutectic freezing period is reached. Yet the beginning of the pro-eutectic freezing is as truly the beginning of freezing in one case as in the other. The difference is not in essence but only in outward appearance. We shall find a closely parallel difference between hypo- and hyper-eutectic cast iron, a difference which leads to confusion among those who will not look beyond the surface.

162. The quantity of the pro-eutectic component and the length of the pro-eutectic freezing period increase with the initial deviation of the salt content from the eutectic ratio, as is evident from the fact that the pro-eutectic freezing period represents the approach of the mother-liquor to the eutectic composition. When the solution is of the eutectic composition to start with, there is no pro-eutectic period, no freezing before the temperature sinks to $-18.5°$, but at this temperature the whole freezes as the eutectic without selection. When the quantity of the salt present is zero, *i.e.*, in the case of pure water, the whole of the freezing is pro-eutectic in a sense, in that it occurs wholly at the zero ordinate and at 0°, in short at point *A*, so that here the eutectic freezing period is zero as it was 100 per cent. in the prior case.

CHAPTER 8

INTRODUCTION TO THE CARBON-IRON DIAGRAM (CEMENTITE-AUSTENITE OR METASTABLE FORM)

(A) Solidification

163. Summary.—The carbon-iron equilibrium diagram, Fig. 23, or map showing the behavior of the carbon-iron alloys, steel and cast iron, with changes of temperature, is of two like parts, the solidification range and the transformation range. Each of these is represented by a group of three lines forming an underscored V, *ABD*, *AEBC* for solidification, and *GOSE*, *PSK* for transformation. This chapter considers only the upper group representing solidification. For simplicity it considers only pure carbon-iron alloys, free from impurities. The formation of graphite is deferred to Chapter 11, §287.

A cast iron with 4.30 per cent. of carbon, *Bn''* Fig. 23, is called eutectic cast iron. Alloys with less carbon, both steel and cast iron, are called hypo-eutectic alloys, and those with more are called hyper-eutectic cast iron.

A eutectic cast iron solidifies unselectively at a single temperature, 1,135°, to form the solid eutectic, ledeburite, a honeycomb of eutectic cementite filled with eutectic austenite, which is gamma iron saturated with cementite (?) dissolved in it. In cooling, the eutectic cementite remains unchanged, but the eutectic austenite transforms progressively into a mixture of cementite, called pro-eutectoid, with pearlite, and this mixture is called eutectic austenoid.

The solidification of hypo-eutectic alloys, white cast irons and steels alike, begins with a pro-eutectic or selective period, in which primary austenite, that is, gamma iron containing cementite dissolved in it, solidifies out progressively, thus giving rise to a mushy stage, with consequent enrichment of the residual molten metal in carbon. If, as happens with alloys containing 1.70 per cent. of carbon or more, and may happen with those containing a little less, this enrichment brings the residual molten up to 4.30 per cent. of carbon, this residue then freezes unselectively and at constant temperature, 1,135°, as the eutectic, ledeburite. Hence, in such alloys there are two distinct periods of solidification: the selective pro-eutectic period with falling temperature, in which primary austenite solidifies, and the unselective eutectic period at constant temperature, in which the remainder solidifies as eutectic.

In still poorer alloys, including nearly all the industrial steels, this enrichment does not usually reach the 4.30 per cent. or eutectic ratio, and here there is only a single stage of solidification, in which austenite progressively richer in carbon solidifies, forming grains or allotriomorphic crystals, the

central part of each of which is poorer in carbon than its outer and later solidified part. This solidification too is selective and at falling temperature.

In hyper-eutectic cast irons, according to the simplified hypothesis adopted for illustration, the excess of carbon over 4.30 per cent. solidifies out progressively as cementite, called primary, and thus impoverishes the residual molten metal till its carbon content falls to the eutectic percentage, 4.30, when this residue freezes unselectively and at constant temperature, 1,135°, as the eutectic ledeburite.

The liquidus, *ABD*, is the locus of the temperature and carbon content of the molten metal during solidification, and the solidus *AEBC* is the locus of the solid matter which is deposited from that molten, primary austenite of primary cementite during the pro-eutectic part of the solidification, the eutectic ledeburite during the eutectic part.

In the slow cooling of the solidified metal, the primary cementite of hyper-eutectic cast iron remains unchanged, but the primary austenite of hypo-eutectic alloys, and the austenite of the eutectic, whenever any is formed, undergo the transformation described above for the eutectic austenite of a eutectic cast iron. When thus transformed the primary austenite may be called primaustenoid. In a rapid cooling this transformation is caught in transit, in part in the intermediate stage of martensite, as in the hardening of steel.

164. Introduction.—Having seen in Chapters 5 and 6 what the constituents of iron and steel are, let us in this and the following chapters consider how they come into existence. As he is a reasonable man who is not deterred from using a hand truck for moving his trunk by the consideration that he has thus to move a trunk plus a truck, so am I reasonable in putting into your hands this useful, indeed indispensable, tool for mastering the A B C of iron metallurgy.

This present chapter considers the solidification of iron and steel; Chapter 9 (p. 158) takes up the transformations as they occur in steel; Chapter 10 (p. 205) studies their occurrence in cast iron and closes (p. 211) with a summary of this study of the diagram as a whole; Chapter 11 surveys the graphitization of the cementite of cast iron (p. 215); and Chapter 12 (p. 231) explains the phase rule, Willard Gibbs's brilliant generalization of the laws governing these diagrams.

The reader unfamiliar with such diagrams as this is cautioned again that he should master the preceding chapter on the sodium-nitrate diagram before taking up the present chapter, proceeding as ever through the simple to the complex.

165. Under what Circumstances and How do the Carbon and Iron Change from One State or Combination to Another? The Carbon-iron Equilibrium Diagram.—We have seen in a general way that, apart from the unchanging primary and eutectic cementite, austenite is the form normally assumed at high temperatures, (§95) and pearlite with ferrite or cementite at low temperatures (§70). Let us now look more closely into the mechanism of the changes

from one state to the other, and the conditions which govern them, subjects which are evidently of great practical importance because of the very profound influence which these different states have on the properties of the metal. Indeed our methods of heat treatment, our hardenings, annealings, temperings, grain refinings, and fibrings, consist in large part in bringing the iron and the carbon to states which will give the metal as a whole the special properties which fit it for the service in view.

The temperatures at which these changes occur, and to a certain extent the nature of the changes themselves, vary very greatly with the carbon content of the metal, and these variations are most readily made clear to the eye by a sort of map, called the Carbon-iron Equilibrium Diagram, Fig. 23, p. 130.

166. Each Ordinate Traces the Course of a Given Steel or Cast Iron.—To explain, every isocarb or vertical line, actual or imaginary, in this diagram represents the condition of a steel or cast iron of a certain given carbon content, at all temperatures covered by the diagram. Thus, following the ordinate of 0.90 per cent. of carbon, the fact that between 0° and 725° it is in Region 8, the pearlite region, represents the fact that a steel of 0.90 per cent. of carbon when in this range of temperature normally consists of pearlite; and the fact that on rising above 725° it enters Region 4, the region of austenite, means that at all temperatures above 725° it normally consists of austenite.

Of course, if at any temperature the mass is composed of two distinct constituents, if for instance a 4.30 per cent. carbon cast iron consists of ferrite and cementite, the composition of each is represented by its own ordinate, the ferrite, because it is free from carbon, by the axis AN, and the cementite, because it contains 6.67 per cent. of carbon, by the axis LD. But, as this cast iron taken as a whole retains at all temperatures this 4.30 per cent. carbon content, so it is represented at every temperature by the point on $Bn'n''$ horizontally opposite that temperature. The reader should have this meaning of the diagram perfectly clear in his mind before going farther.

At every temperature at which any given ordinate crosses a boundary line of one of these eight regions into which the diagram of Fig. 23 is divided, the steel or cast iron which it represents ought to change its constitution from that of the region now left to that of the region now entered. The illustrations which will be given in §§170 to 276 will explain this diagram in greater detail.

To be accurate, because it is an *equilibrium* diagram, this diagram tells us what this constitution and physical state would be if the metal were in equilibrium, *i.e.*, if it had passed into the state which it tends ultimately to assume at the temperature in question, in spite of the sluggishness and *lag* (§222) of these changes, especially of those which occur in the transformation range.

167. The Stable and the Metastable Diagrams.—In fact two such maps are needed, because even for one and the same carbon content and one and

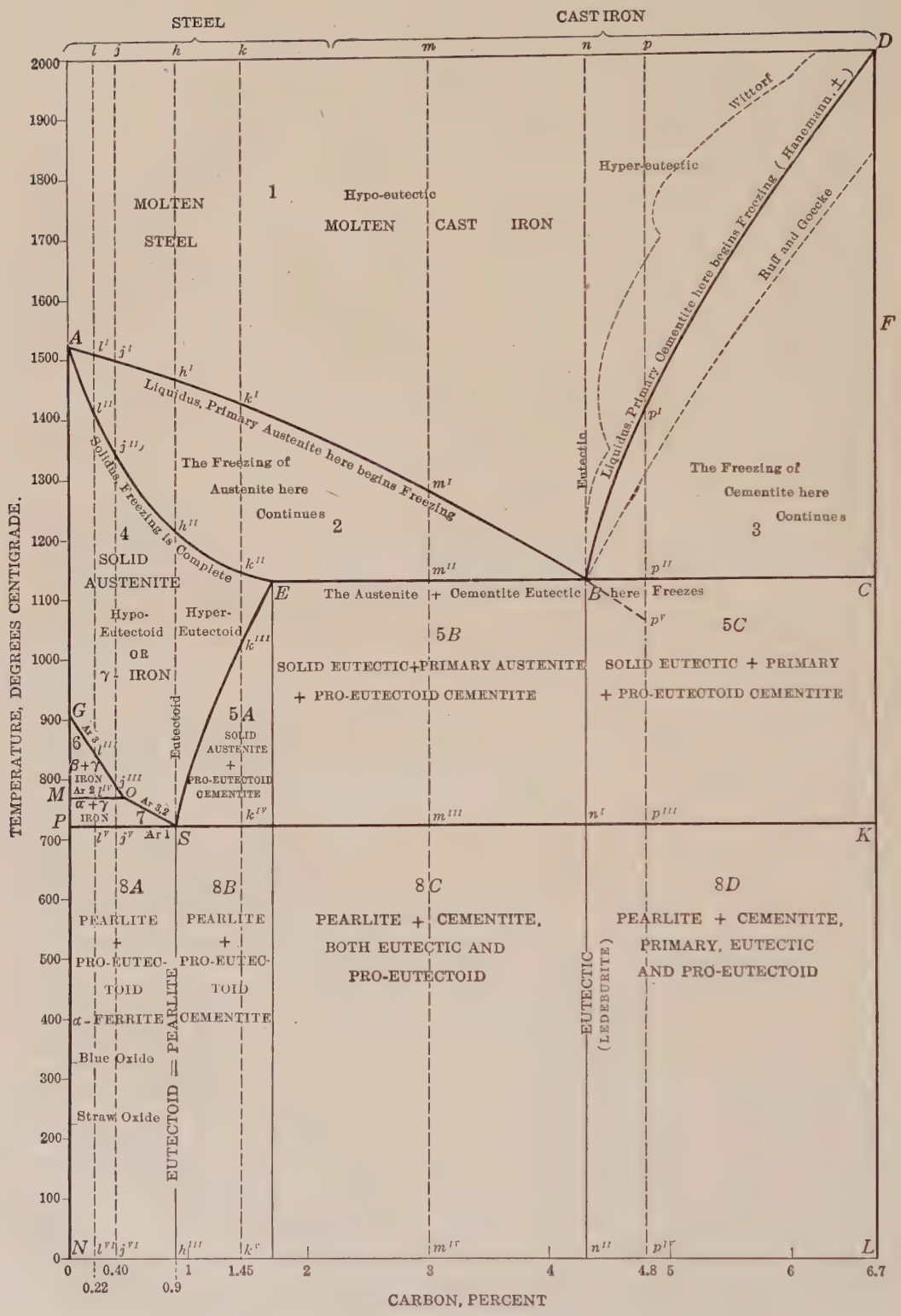


FIG. 23.—The metastable or cementite-austenite carbon-iron diagram.

the same temperature the carbon will, under certain conditions, pass into the state of cementite and under others into that of graphite, as is familiar to us all in the case of white and gray cast iron respectively. Hence we have the iron-cementite or metastable diagram and the iron-graphite or stable diagram, the terms "metastable" and "stable" referring to the fact that the condition of graphite is the more stable of the two, and that of cementite the less stable or "metastable," always tending to shift over into that of graphite, though in case of most steel this tendency is in general held in check by what may be likened to molecular friction or inertia.

Our chief reason for taking up the metastable diagram before the stable one is that it shows us the changes which actually occur in the mechanical and heat treatment of steel, and the temperatures at which they occur, and that the topography of this diagram has been far better explored than that of the graphite-iron diagram.

168. The Cementite-austenite or Metastable Diagram, Fig. 23.[1]—Its general divisions are:

[1] Figure 23 shows only those lines of this diagram which are necessary to a general understanding of the metallurgy and metallography of iron. Other lines are reported, some of which will be considered later. For instance, Carpenter and Keeling reported a continuation of the line *MO* across the diagram, and this continuation is found at times and apparently capriciously in the observations of others. More than one line below *Ar*1 has been reported. For late notes on these lines see Robin (*Iron and Steel Inst.*, Carnegie Schol. Mem., 1911, vol. 3, p. 124, with a note by Osmond on p. 214, and *Proc. Sixth Congress, Internat. Assoc. Testing Materials*, New York, 1912, II, 5), on the transformations which occur below *Ar*1, and Ruer and Klesper on transformations above 1,135°, and their bibliography (*Ferrum*, June, 1914, vol. 11, p. 257).

Note to Fig. 23.—The determinations on which the diagram as here drawn is actually based are as follows:

A, Burgess (G. K.) and Waltenberg (*Bull. Bureau of Standards*, Washington, 1913, vol. 10, Reprint No. 205, p. 8) give four observations in an Arsem furnace made *in vacuo* varying from 1527° to 1531°, mean 1529½°, and two made in hydrogen at 1523° and 1527°, mean 1525°. Eleven determinations on electrolytic iron (p. 7) by the micro-pyrometer method gave values from 1524° to 1538°, mean 1533°. From his various data Dr. Burgess takes 1530° as the most probable temperature for the melting point of pure iron. (Private communication, April 17, 1914.)

AB, Carpenter and Keeling (*Journ. Iron and Steel Inst.*, 1904, No. I, vol. 65, p. 224). Their results are here raised slightly, so as to make the line end at Burgess' point A. The discrepancy between the results of Carpenter and Keeling and the 1530° position of A found by Burgess is less than it might seem at first. The former are on the thermo-electric scale, which Carpenter finds to be about 20° below the optical scale at this temperature. (*Journ. Iron and Steel Inst.*, 1908, No. III, vol. 78, p. 290.) This latter Burgess finds closer than the thermo-electric to the true gas-thermometer scale. Making these allowances the discrepancy is only some 10°C.

AE, *ES*, and *EBC*, Gutowsky (*Metallurgie*, 1909, vol. 6, p. 731).

GOS Howe and Levy (*Trans. Amer. Inst. Mining Eng.*, 1913, vol. 47, p. 587). See also the Author, "Discussion of the Existing Data as to the Position of Ae3," *ibid.*, p. 611; *Stahl u. Eisen*, 1914, vol. 34, pp. 593 and 1142; and *Rev. Metallurgie, Extraits*, 1914, vol. 11, pp. 94 and 98.

S, The Author (*Trans. Amer. Inst. Mining Eng.*, 1913, vol. 47, p. 740; *Rev. Metallurgie, Extraits*, 1914, vol. 11, p. 102).

BD, Hanemann (*Stahl u. Eisen*, 1911, vol. 31, p. 333, and *Zeits. anorg. Chem.*, 1913–1914, vol. 84, p. 1).

Ruff and Goecke (*Metallurgie*, 1911, vol. 8, p. 417).

Wittorf (*Journ. Russ. Physico-chemical Soc.*, 1911, vol. 43, p. 1613; *Rev. Metallurgie, Mem.*, 1912, vol. 9, p. 600; *Zeits. anorg. Chem.*, 1912, vol. 79, p. 1; Stoughton, *Trans. Amer. Inst. Min. Eng.*, 1912, vol. 44, p. 478.

I. *The solidification range*, Regions 2 and 3 of Fig. 23, bounded by the underscored V group of lines, *ABD*, *AEBC*.

II. *The transformation range*, Regions 5, 6, and 7, often called *the critical range*,[1] bounded by a like underscored V group of lines, *GOSEC*, *PSK*.

Of these (I) corresponds closely to the solidification or freezing-point diagram of sodium nitrate water, Fig. 20, §132, to which I shall refer very frequently in order to make the correspondence clear; and (II), closely similar to (I) in its shape, represents a very remarkable group of transformations which occur within the solid metal. In addition there are:

The austenite range, Region 4, which lies between the freezing-point and the transformation range, and the *sub-transformation* or *pearlitic* range, Regions 8*A*, 8*B*, 8*C*, and 8*D*, which extends from the transformation range down to and below the atmospheric temperature.

It is chiefly by playing on the transformations of the transformation range, and on the structural arrangement of their products, that we give our various classes of steel the qualities which fit them to their special uses. Indeed, (I) might be called the important range for the metallurgy of cast iron and (II) for that of steel; and accordingly (II) will receive most of our attention. But for simplicity of exposition we will first take up (I), studying the behavior first of cast iron and then of steel during solidification.

169. Other Illustrations of Solidification and the Genesis of the Various Constituents.—This genesis is shown quantitatively in a series of six additional diagrams, I to VI in the lower part of Fig. 24. In order to show their relation to the carbon-iron diagram of Fig. 23, a miniature of it is set above them. Each of these six diagrams represents this genesis for one of the cast irons and steels which have been selected for the purpose of illustrating the carbon-iron equilibrium diagram, Fig. 23. Their meaning will be explained in connection with these various examples.

The six diagrams of Fig. 25 illustrate graphically the progress of solidification in another way, and the nine diagrams of Fig. 30, p. 163, after Stead, show graphically the structural changes which constitute the transformation in nine steels varying in carbon content from 0.12 to 1.70 per cent.

These various diagrams are lettered to correspond to each other. The actual microstructural changes with varying temperature are illustrated in Plates 8 to 13. Here we may again consult the series of micrographs, Figs. A to O of Plate 3, which shows the progressive change in the microstructure of slowly cooled steel and of white cast iron as the carbon content rises from 0.02 per cent. to about 4.50 per cent.

Apart from these actual micrographs, the general diagrams, Figs. 23, 24, 25, and 30, give us three independent ways of following the changes in iron and steel with varying temperature and varying carbon content

[1] The name "transformation range" is used in this work in preference to the commoner name "critical range" as being much more indicative, and as accenting the solid transformations as distinguished from solidification. "Critical" is very vague. Indeed the processes of solidification are every whit as critical as those of transformation, and for that matter every range in which anything notable occurs is to that extent critical.

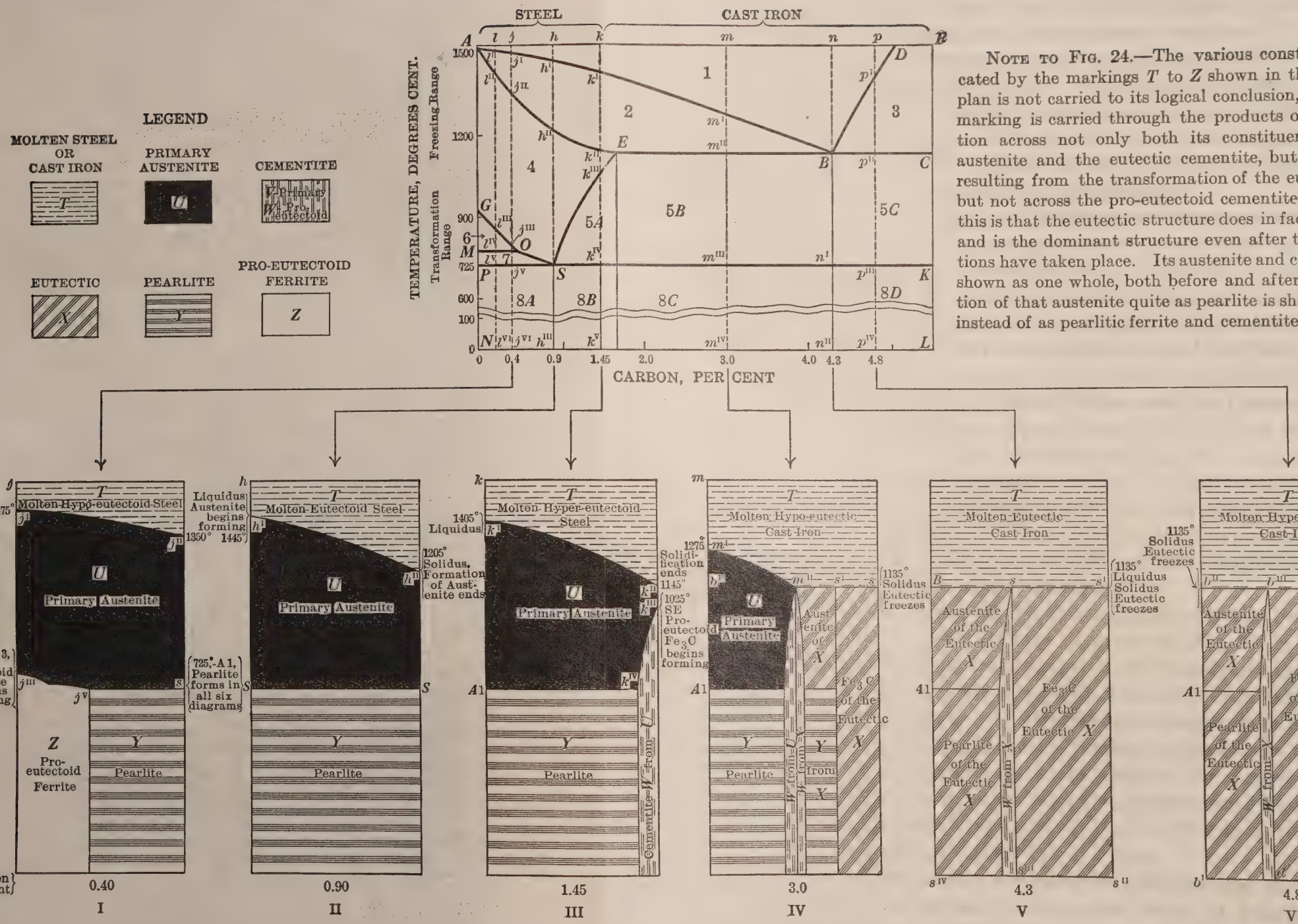


FIG. 24.—Quantitative course of solidification.

NOTE TO FIG. 24.—The various constituents are indicated by the markings *T* to *Z* shown in the legend. This plan is not carried to its logical conclusion, but the eutectic marking is carried through the products of its transformation across not only both its constituents, the eutectic austenite and the eutectic cementite, but also the pearlite resulting from the transformation of the eutectic austenite, but not across the pro-eutectoid cementite. The reason for this is that the eutectic structure does in fact usually persist, and is the dominant structure even after these transformations have taken place. Its austenite and cementite are here shown as one whole, both before and after the transformation of that austenite quite as pearlite is shown as one whole instead of as pearlitic ferrite and cementite separately.

(*Facing page 133*)

thermally, quantitatively, and structurally. It is well to study them all. Fig. 25 will be taken up in the present chapter, because it has to do with our present subject of solidification, whereas Fig. 30, treating of transformation, will be studied in Chapter 9 which is devoted to that subject.

Turning from this general consideration, let us now consider in detail certain specific cases.

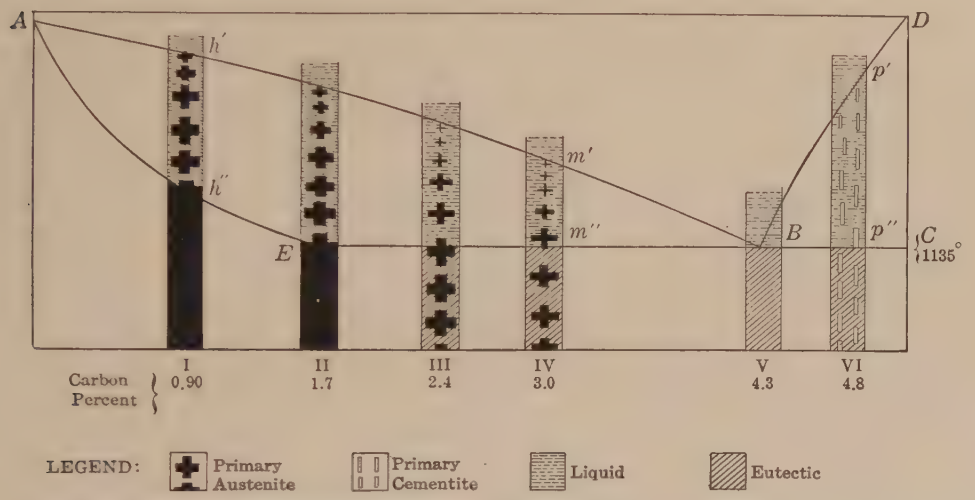


FIG. 25.—Sketches showing the progress of solidification in various steels and cast irons.

170. The Behavior of Cast Iron During Solidification: Cast Iron of 4.30 Per Cent. of Carbon, or Eutectic Cast Iron, Ordinate Bn′n″ of Figs. 23 and 24. —Just as we saw that a eutectic sodium-nitrate solution, *i.e.*, one containing 36.9 per cent. of nitrate, solidifies unselectively at a constant temperature, −18.5°C., and in doing so becomes a mixture of fine particles of pure ice with others of pure nitrate, so a pure[1] cast iron containing 4.30 per cent. of carbon as it cools, or slides down ordinate *nn″* of Fig. 23, solidifies unselectively at a constant temperature, 1,135°, *B*, Fig. 23, and in doing so becomes a mixture of fine particles of two substances, of which one, iron here called "austenite" (§95), corresponds to our pure ice, and the other, cementite, to our pure nitrate. Such a cast iron is a eutectic as we shall see more fully in §172. Its structure is shown in Figs. F of Plate 1, and M of Plate 3, in which the white masses are a honeycomb of eutectic cementite with a dark filling of what at the time of solidification is eutectic austenite. The eutectic itself is the conglomerate or mechanical mixture of these two, the white cementite and the black austenite.

In these micrographs the austenite is in fact replaced by "austenoid" (§80), the mixture of pro-eutectoid cementite and pearlite into which austenite transforms in slow cooling, and this is true of most micrographs representing the eutectic. Had the austenite been retained, had this trans-

[1] A *pure* cast iron, because the presence of impurities shifts the carbon content of the eutectic, as explained in §84.

formation been prevented, the micrograph would have looked essentially as it now does. Hence such micrographs are often and excusably spoken of as showing austenite, though in fact they show only austenoid. (§96, footnote.) This specific eutectic is often called "*Ledeburite,*" especially in Germany.

Cementite (§70) is the definite iron carbide, Fe_3C, which corresponds closely to the definite sodium nitrate particles of our nitrate-water eutectic of §131. The austenite (§95) from which the austenoid seen in our micrographs has been derived is iron saturated with carbon, that is, a saturated solution of cementite in iron. It thus corresponds to our pure ice less closely than the cementite corresponds to our nitrate, but this difference is more apparent than real, because even the pure ice may be regarded as a saturated solid solution of nitrate in ice, a negligibly small quantity of nitrate sufficing to saturate the ice. The behavior of the eutectic in further cooling we will take up in §267.

171. The eutectic here, as in the case of our nitrate-water series, is the alloy of the lowest freezing- and melting-point, *i.e.*, it is the most fusible alloy among the pure iron-carbon compounds, though the freezing- and melting-point may be lowered still further by the presence of impurities, notably by that of phosphorus.

172. The Quantitative Progress of Solidification. Explanation of Diagrams I to VI of Fig. 24.—In each of these the abscissæ of the various curves measure the percentage of the substances present at the various temperatures. For instance, in Diagram I, the abscissæ of the solidification curve $j^{I}j^{II}$ measure the quantity of solid primary austenite which has solidified out of the molten at any given temperature. At 1,475° its abscissa is 0, meaning that at this temperature the quantity of austenite is still 0, and that of the remaining molten is by difference 100, and that the solidification of austenite is about to begin. At j^{II} or 1,350° its abscissa is 100, meaning that, in cooling from j^{I} to j^{II}, 1,475° to 1,350°, the quantity of primary austenite which has solidified is 100 per cent., so that the whole of the molten has solidified as austenite.

The fact that the line $j^{I}j^{II}$ is curved, convex upward, means that as solidification proceeds, the quantity of solid forming in each degree fall of temperature decreases.

The fact that there is no line in this diagram between 1,350°, j^{II}, and 805°, j^{III}, means that the austenite which solidified between j^{I} and j^{II} undergoes no change represented by this diagram, in its further cooling, till it reaches 805°.

The inclined line $j^{III}j^{V}$ means that, in cooling from j^{III} or 805° to j^{V} or 725° there is a progressive generation and separation of pro-eutectoid ferrite out of austenite as the temperature falls. Its being convex upward means that, as this generation of ferrite proceeds, the quantity generated for each degree fall of temperature decreases, and the break between j^{III} and j^{V} means that the rate of this decrease here changes.

Thus an inclined line means a progressive generation of the substance

shown below the line out of that shown above it, with falling temperature, and a horizontal line shows such a generation at constant temperature.

Returning to our eutectic cast iron of 4.30 per cent. of carbon, and applying what has just been said to Diagram V of Fig. 24, the fact that the line Bs^{I}, which separates the molten above 1,135° from the solid below it, is horizontal means that the cast iron solidifies, not progressively as the temperature falls, but all at one constant temperature, yielding the relative proportions of eutectic austenite and eutectic cementite represented by the length of Bs and ss^{I} respectively. The two columns, $Bss^{III}s^{IV}$, and $ss^{I}s^{II}ss^{III}$ represent these solid products in their further cooling, the lines which cross the former representing the transformation which it undergoes in further cooling, as will be explained in considering the transformations.

173. Sketch Showing the Mechanism of Solidification.—Diagram V of Fig. 25, indicates this solidification of the 4.30 per cent. carbon or eutectic cast iron at one fixed temperature, 1,135°, by showing all the space above 1,135° as molten and all below it as solid.[1]

174. The mode of crystalline growth during solidification will be considered later rather fully, but we may now take a superficial glance at it in order to be able to interpret the micrographs used for illustrating the remainder of this chapter, which show the structure developed by solidification.

Fig. 26.—Stages in the solidification of a non-eutectiferous alloy or a pure metal.

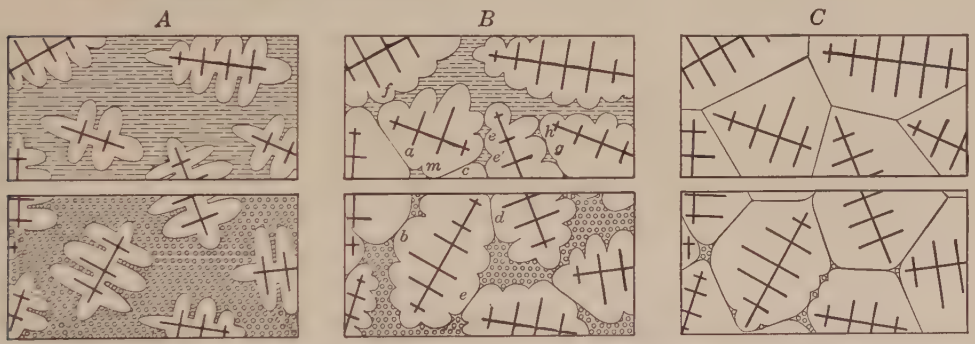


Fig. 27.—Correspondence between the structure of eutectiferous alloys and the stages of Fig. 26.

The solidification of a pure metal may for simplicity be taken as starting as a pine-tree or dendritic outgrowth, nearly simultaneously from a number of independent centers, and extending in every direction from each of them, somewhat as sketched in Stage *A* of Fig. 26. Note that the upper half of the three sketches, *A*, *B* and *C*, belongs to Fig. 26, and the lower half to Fig. 27. In Stage *A* each island represents the dendritic mass of metal which has already solidified about a given nucleus, its crystalline axis being sketched as a sort of back bone. It is now a true crystal, or rather an aggregation of crys-

[1] The representation is not strictly conventional. The very fact that the column has any width indicates strictly that it includes carbon contents greater and less than 4.30 per cent. But let us allow for this and consider that here, as in Fig. 30, each column represents the carbon content which is axial to it.

tals, all oriented alike, with a shore which, though mossy, is a moss composed of the tips of true pine-tree crystals.[1] These, on enormous magnification, would look like the down-growing austenite crystals in the frontispiece.

At a somewhat later stage, *B*, the shores of these islands have spread so far that they have come into contact with each other here and there, as at *a*, *c*, *d*, and *e*, and at these contacts there can be no further growth. But note carefully that the smallest of the landlocked patches of molten now remaining are concave, with the solid projecting convexly into them, suggesting that the solid is male to the molten as female. Here they form concave sided or spandrel-like stars, three-rayed as at *g* and *h* when the end of a bough from one island is opposite the crotch between boughs on its neighbor, and four-rayed as at *e*, *e'*, and *f* when bough meets bough. There they form variously shaped concave masses as at *m*.

In the present case of a pure metal these spandril or male-and-female outlines which exists at this stage cannot be traced in the fully solidified metal, because in the further solidification, from Stage *B* to Stage *C*, each of the particles of the now concave molten bodies adopts, in the act of solidifying, the orientation of the island against which it solidifies, and becomes completely identified with it.

Thus each star or spandrel shrinks progressively during solidification, and is at last reduced to a mere band of demarcation between abutting and differently oriented islands, though as we shall see this band is probably never reduced to a mathematical line.

The islands themselves in the fully solidified metal, *C*, are irregular allotriomorphic polyhedral crystals or grains, each of which has prevented its neighbors, and has been prevented by those neighbors, from retaining the dendritic or idiomorphic outline which it had during Stage *A*. The sections of these polyhedra are the familiar polygons which we see in Stage *C*.

But though their outward form is thus destroyed by their reciprocal interference, their interior symmetry is retained perfectly. Each of these polyhedra has its own orientation, and is in every respect save its outward form a true crystal.[2] If this is clearly understood there can be little harm in the

[1] It would carry us too far here to discuss the apparent discrepancy between the sharpness of the pine-tree shape of the growing crystals of austenite as shown in the frontispiece, and the smoothness of the actual contacts which we find in the solid metal. This sharp pine-tree outline, if adhered to, might be expected to result in a saw-tooth boundary between adjacent austenite grains, and between the primary austenite and the eutectic, instead of the smooth boundaries shown in Stage *C* of Fig. 26 and in Fig. 27. But both kinds of boundaries are in fact smooth. Those between austenite grains are nearly straight lines in our microsections, and those between austenite and eutectic are nearly circular and of short radius, as if surface tension, toward the close of solidification, had smoothed the pine-tree ends into spheroids, and the expected saw-tooth boundaries into planes. It is instructive that the upper sides of the lateral growths of the great stalactite, Fig. *E* of Plate 2, are much sharper than the lower sides, and that little spines often protrude upwards from them. The dulling of the lower sides may be referred to gravity.

[2] See the Author, "What is the Essence of Crystalhood?" *The Metallographist*, 1902, vol. 5, p. 52.

common custom of calling them "grains," to distinguish them from the idiomorphic crystals which sometimes form.

Such grains are seen in Fig. *C* of Plate 21, of austenite in Hadfield's manganese steel; A of Plate 3 of ferrite in pure electrolytic iron; T of Plate 9, of ferrite in steel of 0.21 per cent. of carbon; and A of Plate 10, of austenite in a steel of 1.60 per cent. of carbon, here transformed into martensite. In the second and third of these the grains happen to result from transformation and not directly from solidification.

175. Eutectic Colonies.—The solidification of the eutectic is not quite so simple, because its two components, its austenite and its cementite, solidify and crystallize simultaneously. Nevertheless their joint solidification does follow a generally like plan. For instance in Fig. G of Plate 1 the eutectic occurs in colonies, each with its own general orientation, and the colonies have evidently grown out each from its own nucleus or axis till they have met. The dark irregular band which separates neighboring colonies appears to be the last solidified part of the eutectic austenite which has been drawn together by surface tension. Their shape, female to the main masses of the colonies as male, testifies to their being the last to solidify, occupying the space left by the solidification of the earlier parts.

Turning now to the pure eutectic shown in F of Plate 1 and M of Plate 3, a like colonial habit can be traced. In some places the plane of the section has cut the honeycomb across and in other places along the length of its cells, giving rise in the former to black austenite spots in a white cementite matrix, and in the latter to black and white zebra stripes. Each such spotted region represents one eutectic colony, and each striped region another.

176. The Solidification of Hypo-eutectic Cast Iron of 3 Per Cent. of Carbon, Ordinate m-m^{iv} of Figs. 23 and 24.—This solidification differs from the preceding case, first in that it is divisible into two distinct parts, a pro-eutectic or primary part which is spread out over some 140°, from about 1,275° to about 1,135°, and a eutectic part which takes place at a fixed temperature, about 1,135°; and second in that the pro-eutectic part is selective, the solid layers deposited at each instant being poorer in carbon than the molten whence they deposit, with the consequent progressive enrichment of that molten in carbon.

During the passage down through this solidification range from 1,275° to 1,135° the metal is in a mushy stage (§182) like that of the freezing sea water so familiar to every shore-bred boy, or like sherbet. At any given instant so much of the metal as is in this range of temperature is an intimate mechanical mixture of certain particles which are distinctly solid, like the minute particles of ice or snow in our sherbet, and others which are distinctly liquid like the water which is between those little particles of ice.

In the solidification of a large mass it usually happens that the cooler outer parts first to solidify pass down through and below this mushy range while the hotter interior parts later to solidify are still in that range, so that they form two distinct regions, a distinctly solid continent, and these shore

regions which are still in the mushy stage. Under these conditions the mushy shore layer forms a sort of mossy growth, composed of minute dendritic or tree-shaped crystalline masses, between the trunks and boughs of which lies the still molten mother-metal.

177. The Pro-eutectic Solidification.—Just as our hypo-eutectic nitrate solution, containing less than the eutectic percentage of nitrate, (1) begins freezing when, in Fig. 20, its temperature has sunk to $-4°$ on its liquidus AB; and just as this part of its freezing consists (2) of the progressive crystallizing out of so much of the water as is in excess of the eutectic percentage of water, 63.1, and (3) the consequent progressive enrichment of the solution in nitrate, at such a rate that the nitrate content of the solution at each temperature is measured by the abscissa of AB for that temperature; so this poorer-than-the-eutectic, *i.e.*, this *hypo-eutectic* cast iron, which, we may imagine, has been preheated to m or 2,000°, (1) begins to solidify when, in cooling, its temperature reaches m' or about 1,275° on the liquidus AB of Figs. 23 and 24. (2) This part of its solidification is of crystals of the excess or primary substance, iron here called austenite; causing (3) the enrichment in carbon of the molten mother mass at such a rate that at each temperature below m' its carbon content is measured by the abscissa of AB, which is the *locus of the molten, i.e., the liquidus.* This conception that the liquidus represents the history of the molten mother metal, its temperature and carbon content at each step in the progress of solidification, should be prominently in mind.

Though this solidification of primary austenite is truly selective, the selection is incomplete, as is explained in §203.

178. The Eutectic Solidification.—And just as, once the nitrate content of the aqueous solution has been thus enriched up to the eutectic content of 36.9 per cent., the further freezing is of the eutectic, unselectively and at the constant eutectic temperature of $-18.5°$; so the further solidification of this cast iron, once its molten part has been thus enriched up to the eutectic carbon content of 4.30 per cent., is that of the eutectic, ledeburite, unselectively, at the constant eutectic temperature of 1,135°, at which the whole solidification of the eutectic cast iron of §172 occurred.

Just as the ice which froze out from a hypo-eutectic nitrate solution before the eutectic, that is in cooling from the liquidus to the solidus, was called "pro-eutectic ice," to distinguish it from the eutectic ice which is one constituent of the eutectic; so this austenite which solidifies during the cooling from m' on the liquidus to m'' on the solidus, and thus before the eutectic, may be called "pro-eutectic austenite," to distinguish it from the austenite which is one constituent of the eutectic ledeburite, that is from the eutectic austenite. But the term "*primary austenite*" is on the whole more acceptable, because it is much shorter and on the whole more indicative. "Pro-eutectic" is unfortunate in sounding much like "pro-eutectoid," which as we have seen has a wholly different meaning (§72).

179. Replacement of the Eutectic by Cementite.—This, the normal or rather the equilibrium course of solidification just described, is often departed

from. The progressive solidification of austenite and consequent enrichment of the molten mother-metal, instead of coming to an end and being succeeded by this solidification of eutectic when that enrichment has reached the eutectic carbon content of 4.30 per cent., may continue by what is known as *undercooling*. Just as during normal solidification the temperature and carbon content of the molten are represented jointly by the liquidus, in the present case of a 3 per cent. carbon cast iron by the part of the liquidus from m' to B, so during this prolongation of the solidification of austenite past the point B at which it ought to cease, they are represented by a continuation BB', of AB. To fix our ideas we may consider a case in which undercooling has thus brought the molten to point p^{v}, Fig. 23, and to the same end we may adopt BD as the hyper-eutectic liquidus.

Any such undercooling is out of equilibrium and in that sense unstable. If the freezing of water is prevented from occurring when the temperature reaches and falls below 0°, the instability of the water thus remaining liquid increases as the temperature falls further below 0°, and finally becomes so great that the tendency to freeze becomes irresistible. We may say that the degree of instability of undercooled water is represented by the temperature gap between its existing temperature and 0°, at which freezing is due.

But the upper limit of this temperature gap in the case of an undercooled molten or aqueous solution, say of our 3 per cent. carbon cast iron undercooled to p^{v}, is not p^{II} at which the solidification of austenite ought to have ceased, but p^{I} on the liquidus, as the least reflection makes clear. For if there were a molten containing p^{v} or 4.80 per cent. of carbon heated hotter than p^{I} and isolated within a magnesia crucible, and if this were then cooled to the liquidus, p^{I}, the precipitation of cementite from it would then be due, by the definition of the liquidus. But if the molten from our 3.80 per cent. carbon cast iron has been enriched by undercooling to p^{v} or 4.80 per cent. of carbon, the temperature gap by which it is below the point p^{I} at which, in view of its carbon content, it ought to deposit cementite, is $p^{v}p^{I}$, and this may be taken as the measure of its instability. When, as in our present case, the right-hand liquidus is much steeper than the left-hand one, a small degree of undercooling below the eutectic temperature represents a great temperature gap of instability.

In such undercoolings there is often a degree of undercooling which cannot be exceeded, the limit between the metastable undercooling and the labile. And in any event a time would have to come at last when the further freezing out of austenite is not simply opposed by the increasing instability, that is the increasing tendency of the cementite to precipitate from the molten which is becoming more and more supersaturated, but necessarily stopped by the exhaustion of the supply of austenite, when the molten has been enriched till its carbon content has reached 6.67 per cent. and it is now all cementite. And the undercooling is likely to end before the enrichment proceeds so far. It is followed by the solidification of cementite, and thus by an impoverishment of the abnormally enriched molten metal, carrying

it back toward the eutectic carbon content of 4.30 per cent. On reaching this carbon content, or one related to it, the eutectic, or a body related to it, at last solidifies. But there will be less of the eutectic, or rather "hypertectic" then formed than if it had solidified when the molten metal first reached 4.30 per cent., less by the amounts of the austenite and of the cementite which have solidified out since that moment, and it may indeed be nil, at least locally.

This replacement of part of the eutectic cementite is common in the rapid cooling of white cast iron. It is of course more accurate to say that part or all of the eutectic is replaced by austenite and cementite which have solidified in successive stages instead of simultaneously, and hence have not taken on the honeycombed arrangement to which their simultaneous solidification leads. Cases of this replacement are considered in §274.

180. The quantitative progress of solidification is shown in Diagram IV of Fig. 24, as the progressive increase in the quantity of primary austenite from m^{I} or 1,275° to m^{II} or 1,135°, and the solidification of the remainder as eutectic at m^{II}. Here, as in Diagrams I, II, and III of this group, the abscissæ of the curve which bounds the lower side of the molten measure the quantity of primary austenite which has solidified out of the molten at each temperature. Thus in cooling from the liquidus 1,275°, m^{I}, to the solidus 1,135°, m^{II}, the quantity $b^{II}m^{II}$ of primary austenite thus solidifies, enriching the remaining molten up to the eutectic carbon content 4.30 per cent. In now cooling past the eutectic freezing point, 1,135°, the whole of the remaining molten solidifies unselectively, yielding $m^{II}s$ per cent. of eutectic, composed of $m^{II}s^{I}$ of eutectic austenite and $s^{I}s$ of eutectic cementite. These three products, primary austenite, eutectic austenite, and eutectic cementite, are represented by the three columns headed $m^{I}m^{II}$, $m^{II}s^{I}$, and $s^{I}s$ respectively, while the transformations in the austenite, both primary and eutectic, in cooling further are represented by the lines which cross these columns. These will be explained in connection with the transformations.

181. A sketch of the course of crystallization is given in Diagram IV of Fig. 25. Note the progressive increase in the quantity of the black crosses, suggesting the cross-sections of primary austenite pine trees, as we pass down from m^{I} on the liquidus AB to m^{II} on the solidus EBC. Note further this progress of solidification throughout the range $m^{I}m^{II}$ instead of its restriction to a fixed point, B, as in the next diagram, V. Note further that the primary austenite which has thus solidified between m^{I} and m^{II} remains in the solid metal at lower temperatures, but now imbedded in a matrix of eutectic, which itself is identical with the eutectic which forms the whole of the eutectic cast iron of Diagram V. This drawing out of solidification over a range of temperatures does not represent lagging. If the metal were cooled down to half way between m^{I} and m^{II} and held at that temperature indefinitely, only the moderate quantity of primary austenite corresponding to equilibrium would form.

Note finally that, on reaching m^{II}, 1,135°, the whole of the remaining

molten changes at this one temperature into solid eutectic, the progressiveness of the solidification from m^{I} to m^{II} being replaced thus by constant temperature solidification.

182. The mushy stage which thus exists in the range of temperature from the liquidus to the solidus is of great industrial importance. We have seen that during this stage the metal is a mechanical mixture of (*a*) certain particles of primary austenite which have already solidified, have become distinctly a viscous solid, and (*b*) the residual mother-metal which is distinctly molten. When the temperature first descends into this range the quantity of the solid austenite is so small that it easily escapes notice. When it has sunk nearly to the bottom of this range the mushiness is very prominent, because there is now about as much molten as solid.

As the other diagrams of Fig. 25 show, all cast irons except that which is exactly eutectic have such a mushy range, extending from the liquidus to the solidus, and so have all steels except pure carbonless iron, which has not. For all pure carbon cast irons the lower end of this range is fixed, 1,135°, but its upper limit varies with the carbon content, and the width of this range increases as the carbon content departs from the eutectic ratio. I say all pure carbon cast irons, because the presence of an impurity may shift the solidification temperature of the eutectic. For the eutectic-free steels both the upper and the lower limit of the mushy range shift with the carbon content, and the width of this range increases as the carbon content increases from 0 to 1.70 per cent.

183. Burning in the Mushy Stage.[1]—Heating steel well into this range appears to induce "burning," that mechanical deterioration which is so difficult to cure without re-melting. Certainly, to rise far enough into this range to give rise to any considerable quantity of molten particles, within the hitherto viscously solid mass, in itself causes great tenderness, so that the metal cannot be forged or rolled without tearing apart. It is partly because the lower limit, AE, of this range descends so rapidly as the carbon content increases that high-carbon steel is so much more readily burnt, and needs so much lower a forging temperature, than low-carbon steel (§686).

184. Pulling in the Mushy Stage.—Again, it is this mushy stage that leads most readily to the "pulling" of steel and iron castings. During the solidification of these castings the progressive cooling of the metal leads to its contraction, but this contraction is in turn opposed in most cases by the walls of the mould, as the fastening of the ends of a guitar string opposes its attempt to contract when moistened by damp air, with the familiar result that the string pulls itself in two, because of its inability to accommodate itself to its own contraction thus rigidly opposed. To restrain thus forcibly its attempt to pass to the shorter length which its dampened state im-

[1] Stansfield, *Journ. Iron and Steel Inst.*, 1903, No. II, vol. 64, p. 433.

plies is as severe an outrage as it would be to stretch it out forcibly to this greater length after it had been allowed to contract freely. The two things are identical in effect. Such a stretching it is unable to undergo without breaking.

The very fluidity of a fully molten metal enables it to accommodate itself to the contraction which its cooling implies. As long as it is fluid it thus obeys readily its own tendency to contract with falling temperature. And when it has passed below the solidus and become distinctly solid it may be strong enough to crush the walls of the mould, and thus overcome their resistance to its moving freely in obedience to its own contraction. But while it is in this mushy stage, lacking both the freedom of the molten state and the strength of the solid state, it readily tears itself open in attempting to contract while its ends are held apart by the mould.

This difficulty is lessened in gray cast-iron founding by the expansion, during solidification, caused by the sudden graphitization of the cementite.

In the case shown in Fig. D, Plate 6, this pulling apart is shown to have occurred late during the solidification, when the quantity of molten was so far reduced that there was not enough to run down by gravity, yet when there was still enough to convert the rough fracture faces into perfect pine-trees, by its crystalline grouping after the removal of the interfering contact of the adjoining growth.

Fig. 28.—Hannover's spongy tin.[1]

185. The Mushy Stage Utilized.—Hannover[1] takes advantage of this mushy stage to prepare metallic sponges. He heats an alloy, say of lead and tin, above its eutectic melting point, and, rotating it rapidly at this temperature, ejects the molten part centrifugally, leaving a tin sponge representing the part which had not melted, such as is shown in Fig. 28. In this figure the black parts represent the cavities left by the ejection of the molten part of the alloy, and the white parts represent the dendrites of primary tin which remain. In this way he proposes to make spongy lead plates for accumulators, porous silver for holding firmly any enamel set on it, filters, etc.

186. The mode of crystalline growth during the solidification of a hypo-eutectic cast iron may be visualized further by means of Figs. 26 and 27.

[1] From *Comptes rendus*, 1912, vol. 154, p. 1594; and *Revue de Metallurgie, Memoires*, 1912, vol. 9, p. 641.

In Fig. 26, p. 135, the molten, which at Stages *A* and *B* lay as a matrix, or as female to the solid dendritic islands as male, in passing from Stage *B* to Stage *C* has identified itself with those islands, so that no trace of the female or spandril shape is left.

We have hitherto assumed that these sketches represent a pure metal. But Stage *C* of Fig. 26 would result equally in the solidification of primary austenite, or of any other alloy, provided that no eutectic forms. Here, even though the solidification is selective, so that the later solidified layers are richer in carbon than the earlier, this complete identification of even the last solidified particles with the islands against which they solidify must needs occur. It is true that the isocarbs of these islands may retain this dendritic shape, even after the wave of diffusion has gone far toward equalizing the carbon content. But as long as that which is solidifying is primary austenite, each particle as it solidifies adopts the orientation of the island to which it attaches itself, and thus merges with that island. This is as true of the particles which at Stage *B* form the little stars, *e*, *f*, *g*, and *h* as of any other particles. The progressive effacement which they undergo from *A* to *B* continues and completes itself at *C*. Hence, if the whole solidification up to the very end is of primary austenite, it will lead to the polyhedral grain structure which Fig. 26, *C*, shows in section.

But in case a eutectic forms, as is true of this present 3 per cent. carbon cast iron, this identification cannot be complete, but traces of the female to male structure must persist, as sketched in Stages *A*, *B*, and *C*, of Fig. 27. The identification of the depositing layers with those previously deposited must needs end when, at the end of the period of selective solidification, the molten has been enriched up to the eutectic carbon content of 4.30 per cent., for what solidifies henceforth is eutectic, which cannot merge with the existing austenite islands, but instead solidifies as eutectic about those islands, and thus preserves their outline permanently.

Hence results the topographical arrangement which we actually find in such alloys, sketched in Fig. 27, a matrix of eutectic female to the primary austenite islands as male. Here *A*, *B*, and *C* sketch the general topographical relation of the primary austenite to the eutectic, *A* where there is much eutectic, that is when the carbon content falls but little short of 4.30 per cent.; *B* when there is less eutectic, that is when the carbon content is still smaller; and *C* when there is but little eutectic. *A* and *B* of Fig. 27 differ from *A* and *B* of Fig. 26 only in having solid eutectic in place of the molten, whereas *C* of Fig. 27 differs from *C* of Fig. 26 in preserving, in the condition of eutectic masses, the stars and spandrils which in *C* of Fig. 26 have effaced themselves by merging and identifying themselves completely with the earlier solidified mass. Thus *A* and *B* of Fig. 26 might represent stages immediately preceding *A* and *B* respectively of Fig. 27, in Fig. 26 representing the end of the selective solidification, when the residual molten has just reached the eutectic composition, and in Fig. 27 representing the next stage, when the whole of that eutectic has in turn solidified.

187. Illustration of the Crystalline Growth of Hypo-eutectic Cast Iron. A. The Growth of the Primary Austenite During Pro-eutectic Solidification.—The primary austenite, as it progressively solidifies or crystallizes out, is like a crystal, for instance of alum, growing in an aqueous solution, in that it is surrounded by liquid matter on every side, and therefore in growing is free to follow its natural course, and thus to build up idiomorphic crystals, *i.e.*, crystals of the natural form peculiar to the substance itself, as distinguished from allotriomorphic crystals with a shape which the substance is forced by its environment to assume, as when molten iron is poured into a spherical mould, and thus becomes an allotriomorphic sphere, or when a cube of copper oxide, cuprite, is gradually converted into native copper by the loss of its oxygen, and thus gives rise to a pseudomorphic cube of copper.

This idiomorphic shape for austenite is that of an octahedron, habitually a dendritic series of octahedra connected together by what is called "parallel growth," as seen in the frontispiece and in Fig. E, Plate 2. These two probably represent extremely slow solidification. Under the more usual conditions of faster solidification, the austenite is more likely to assemble in branching pine-tree crystals, such as are shown in Figs. A, B, and C, Plate 6 (§§342, 443, 444).

The existence of these free or independent idiomorphic crystals of austenite implies that, at a stage like *A* of Fig. 26, the growing austenite island has been removed bodily from the molten in which it has grown, quite as a specimen crystal of alum in a druggist's window must needs have been removed bodily out of the aqueous solution in which it grew. What is in effect such a removal may come about through the gradual ebbing of the molten interior of a large ingot, because of the varying rates of cooling and hence of contraction in the various concentric layers, resulting in the formation of an axial cavity called the "pipe."

The great elongation by "parallel growth" of the primary austenite crystals shown in the frontispiece represents the progressive ebbing of the molten, which continued feeding their lower ends after the upper part of their sides, left bare by the ebb, was cut off from further growth.

The great Tschernoff crystal of Fig. E, Plate 2, is shown by Prof. Tschernoff's sketch[1] to have grown downward in this way.

Ebbing at such a rate as just to keep pace with the downward growth of the austenite crystals is rare, but a corresponding though much smaller crystalline growth generally lines the sides, and especially the upper side, of any cavity caused by ebbing, in short any contraction cavity.

This crystallization of the primary austenite will be considered more fully in §§772 to 776.

188. B. The Growth During the Eutectic Solidification.—The frontispiece and Fig. E, Plate 2, having illustrated the stage corresponding to Stage *A* of Fig. 26, we may illustrate the stage corresponding to *A* of Fig. 27, the end of the eutectic solidification, by means of Fig. B, Plate 6. This shows the

[1] *The Metallographist*, 1899, vol. 2, p. 74.

microstructure of a white cast iron called "washed metal." It corresponds roughly to *A* of Fig. 27, for the general arrangement which exists at the end of solidification is preserved, the transformations which indeed occur in further cooling giving no evidence of themselves in this micrograph. The structure is not complicated, as it would be in the case of gray cast iron, by the later graphitization of the eutectic cementite.

Here the black bats-wing areas, each with length running NW and SE, are the primary austenite, while the fine black and white mottled or banded ground mass is the eutectic, ledeburite, its white being the eutectic cementite and its black the eutectic austenite.

It is the regularity of arrangement of the black bats-winged areas of Fig. B, Plate 6, that shows that they represent the primary substance, growing freely in a liquid bath. But the plane of our microsection, though it discloses like regularity, may show, not such longitudinal sections of pine-tree boughs, at right angles to the main trunk of the dendrite, but cross-sections of series of boughs, as in Figs. A[1] and C, Plate 6.

The type of growth here shown is the same as that of Fig. B, Plate 6, as is seen on the slightest consideration. For instance, the black bat-wing areas of Fig. B, if cut by a plane running through *ab* and normal to the plane of the paper, would appear as a row of round dots, like those of Figs. A and C, of Plate 6.

This regularity of arrangement is quite enough to identify the primary austenite as having solidified before the matrix in which it lies, without the evidence of the banded structure, which in Fig. B shows distinctly that that matrix is eutectic.

The shape of the individual masses, as distinguished from the regular sequence in which the masses stand, is less trustworthy as a means of identifying the primary substance. For instance, in Figs. A and B, of Plate 11, and G of Plate 13 there are rounded masses of eutectic or of cementite which simulate the shape of many primary austenite masses. Indeed, in cases in which the quantities of primary austenite and of eutectic are nearly equal, neither the shape nor the sequence may suffice to distinguish them. It is enough here to point that when, as in Figs. 27*A* and A, B, and C of Plate 6, the quantity of primary austenite is small, the regular sequence of its masses serves to identify it; and that when, as in Figs. 27*C*, G of Plate 6, A and B of Plate 10, D and E of Plate 12, and E, and F of Plate 13, the quantity of eutectic is small, the spandrel or star shape of its masses serves to identify it. In both cases the black etching of the austenite and the white etching and composite structure of the eutectic are additional signs.

The quantity of eutectic shown in Fig. B of Plate 6 is actually much larger than our present 3 per cent. cast iron would have. The series of micrographs

[1] The reason why the spots in Fig. A are white, whereas the bats-wing areas of Figs. B and C are black, is that the etching for the former was done with sodium picrate and for the latter with picric acid. The regular dendritic mass in each case is certainly the primaustenoid, that is to say the primary austenite, transformed in the slow cooling.

Figs. J to M, Plate 3, shows a progressive increase in the quantity of eutectic with the carbon content of the cast iron.

Note in particular the star or spandrel shape of the eutectic masses in Fig. D of Plate 12, recalling the three and four-rayed eutectic stars sketched in Fig. 27*C*. Just as the regularity of the arrangement of the dendritic masses in Figs. A, B, and C, Plate 6, testified to their being the primary substance, so here the female, spandrel-like, or star-like shape of these masses identifies them as the last solidified part, and hence as probably eutectic, while their banded structure confirms this interpretation.

189. The solidification of hyper-eutectic cast iron of 4.80 per cent. of carbon, ordinate p-p^{IV} of Figs. 23 and 24. *A caution.* Our knowledge of this hyper-eutectic part of the diagram is now evolving, as may be inferred from the presence of three distinct lines representing the hyper-eutectic liquidus, found by three distinguished observers. Their observations will be given in another volume. The purpose of the present chapter is to give a bird's eye view of the diagram, and this is incompatible with either accurate shadings which shall represent the balance of probability between the conflicting data, or repetitions of caution as to the relative probability of this or that phenomenon reported by one observer but contradicted by two others. Fortunately the great mass of industrially important facts agree with the hypothesis as to this part of the diagram which I adopt, a hypothesis moreover which simplifies the diagram very greatly, and in particular facilitates an understanding of the transformation range, the one of chief importance for the metallurgy of steel, by making these two parts of the diagram, the solidification and the transformation ranges, closely parallel. Let us master the subject in its broad simplicity, and later grapple as best we may with its complexities. The inaccuracies which a broad and simple exposition imply in this case are, indeed, matched by corresponding inaccuracies in a like exposition of any other field of natural science. We have only to refine our methods of observation closely enough to find inaccuracies in almost every law of nature simply stated.

190. The Course of Pro-eutectic Solidification.—Just as our hyper-eutectic nitrate solution, containing more than the eutectic percentage of nitrate, (1) begins freezing when, in Fig. 20, p. 112, its temperature has sunk to 40° on its liquidus *BC;* and just as this part of its freezing consists (2) in the crystallizing out of the salt, and (3) the consequent impoverishment of the solution in nitrate, at such a rate that the salt content of the solution at each temperature below 40° is measured by the abscissa of *BC* for that temperature, so this richer than the eutectic, *i.e.*, hyper-eutectic cast iron, which we may imagine has been preheated to p or 2,000°, (1) begins to solidify when in cooling its temperature reaches p^{I} or 1,400° on the liquidus *BC*, Fig. 23; and this part of its solidification consists (2) of the precipitation, within the molten mass, of crystals of the excess or primary substance, Fe_3C, or cementite, and (3) the consequent impoverishment in carbon of the residual molten mother mass at such a rate that its carbon content at each tempera-

ture below 1,400° is measured by the abscissa of *BC* for that temperature, which thus is the *locus of the molten.*

And just as once the salt content of the nitrate solution has by this process been reduced to the eutectic content of 36.9 per cent., the further freezing is that of the eutectic, unselectively and at constant temperature, so the further freezing of this cast iron, once its temperature and composition have reached *B*, is that of the eutectic, ledeburite, unselective and at constant temperature, 1,135°, which is that of the whole of the solidification of the eutectic cast iron of §172.

Just as the austenite which solidified from our hypo-eutectic cast iron of §176 during the cooling from the liquidus to the solidus was called "pro-eutectic" or better "primary austenite," so this cementite which here solidifies during the cooling from the liquidus p^{I} to the solidus p^{II} of our hyper-eutectic cast iron, and hence before the eutectic solidifies, may be called "pro-eutectic" or better "*primary cementite.*" The reasons why "primary" is preferable to "pro-eutectic" in the case of austenite apply here, and are strengthened by the fact that there is a distinct substance, "pro-eutectoid cementite" (§72) which might well be confounded with "pro-eutectic cementite."

191. The Eutectic Solidification.—Just as our nitrate-ice eutectic and the whole of our eutectic cast iron in solidifying break up into their two constituents, ice and nitrate in the former case, and eutectic austenite and eutectic cementite jointly called ledeburite in the latter, so the eutectic which forms during this eutectic freezing period of this present cast iron breaks up into this same ledeburite, a honeycomb of eutectic cementite, with its cells filled with eutectic austenite.

192. The Rigidity of Selection should Make All Hyper-eutectic Cast Irons Eutectiferous.—And just as the purity of the ice freezing out of hypo-eutectic nitrate solutions, its freedom from nitrate, means rigid selection, so that no matter how little salt is present none of it freezes out with the ice but all of it remains unfrozen and concentrates progressively, and at last must needs result in the formation of a quantity of eutectic representing the whole of this salt, so that every such solution when frozen is eutectiferous; so this present freedom of the solidifying cementite from free iron implies a like rigidity of selection, and a like resulting quantity of eutectic ledeburite containing the whole of the iron present in excess of that which can be contained in combination with carbon in the primary cementite under the special conditions of the case. These conditions are that all the iron and the carbon will divide themselves into two bodies of which one, the primary cementite, contains 6.67 per cent. of carbon and 93.33 per cent. of iron, and the other, the eutectic, contains 4.30 per cent. of carbon and 95.70 per cent. of iron. In short, all hyper-eutectic cast irons are eutectiferous on this theory.

In the iron as in the nitrate solution, if the quantity of eutectic is extremely minute, it may be so entangled mechanically that we cannot detect it; but it should be there.

193. The quantitative progress of solidification is shown in Diagram VI of Fig. 24. In order to make clear the correspondence between this Diagram VI and the carbon-iron Diagram itself which stands above it, the abscissæ of the curve $p^{I}p^{II}$ are made to represent the quantity of molten which remains at any given temperature, and the intercept between $p^{I}p^{II}$ and the axis psp^{IV} measures the quantity of primary cementite which has solidified out of the molten at each temperature.

When the temperature falls as far as 1,400°, the liquidus for this carbon content, 4.80, primary cementite begins crystallizing out, and the quantity which has formed by the time when each lower temperature is reached is indicated by the curve $p^{I}p^{II}$. When the temperature reaches the eutectic freezing point, 1,135°, the quantity of primary cementite which has formed is $p^{II}s$. On cooling past 1,135° the remaining molten, represented by the abscissa of p^{II}, solidifies unselectively as the eutectic, represented by the broad zebra striped region composed of two columns, the left-hand one $b^{III}b^{II}b^{I}d$ representing quantitatively the austenite of that eutectic, and the right-hand one $b^{III}p^{II}bd$ representing its cementite.

194. A Sketch of the Progress of Solidification is given in Diagram VI of Fig. 25.—Note the progressively increasing size of the white rectangles representing the primary cementite crystals, as the temperature falls from the liquidus *BD* to the solidus *BC*; the abrupt change of the remaining molten into eutectic at one fixed temperature, 1,135°; and the structure below 1,135°, *i.e.*, after solidification is complete, with these primary cementite crystals imbedded in a ground mass of the eutectic which solidified at 1,135°.

195. Crystallization of Primary Cementite during the Pro-eutectic Solidification.—What has been said in §176, about the crystallization of primary austenite in the corresponding period from hypo-eutectic cast irons applies here, save that the natural form of the primary cementite is that of a rhombohedron, reduced in most cases to a thin plate. In those exceptional cases in which the molten ebbs at such a rate as to keep pace with the growth of these cementite plates, they are found spanning the cavity which that ebb causes. An example of this occurrence is shown in the inset of the frontispiece.

But in the great majority of cases the cementite plates are found simply shooting across the solid ground mass of eutectic, which of course has solidified after them. These cases really differ from that shown in the frontispiece only in that the eutectic has remained in place instead of draining away.

196. The microstructure of such a cast iron is seen in Figs. G, Plate 1, and N and O, Plate 3, in which the great broad white bands are the primary cementite, and the zebra-striped ground mass is the eutectic, identical with the eutectic of Figs. F, Plate 1, and M, Plate 3. The surfaces of these great cementite plates are planes of weakness, along which rupture naturally passes. Hence it is these that form the brilliant faces of "spiegeleisen." (§731.)

The presence of these cementite plates distinguishes hyper-eutectic from

hypo-eutectic cast iron, *i.e.*, from that which is poorer in carbon than the eutectic. They are not commonly seen in foundry cast iron, first because this is not commonly hyper-eutectic, and second because at the high temperature, above 1,135°, at which they form they habitually turn into graphite by the reaction $Fe_3C = 3Fe + Gr$. Their presence in this individual specimen is due to its large content of manganese, which makes the cementite stable. Their presence in Fig. O of Plate 3[1] is due to cooling the metal so rapidly that their graphitization was prevented. This graphitization is considered in Chapter 11.

197. A mushy stage for such hyper-eutectic cast irons is indicated by Diagram VI of Fig. 25. From p^{I} on the liquidus down to p^{II} on the solidus the mass is seen to consist of a mechanical mixture of (1) sheets of white cementite and (2) the residual molten which bathes them. In the case of the manganiferous cast irons, spiegeleisen and ferromanganese, in which this primary cementite is so stable, thanks to its manganese, that it persists, such a mushy stage should be prominent. But it is not seen in hyper-eutectic cast irons relatively free from manganese for a simple reason. These sheets of primary cementite, as fast as they form, turn into graphite, which is so light that it at once floats to the surface and escapes, as explained more fully in §308. This self-removal of the primary cementite through graphitization, emergence, and blowing away, not only thus prevents a true mushy stage from arising, but conceals the true nature of this period, that it is the period of pro-eutectic solidification, so that even very intelligent observers who have not looked at the matter from this point of view may deny that this graphite-discharging period is solidification at all.

198. Solidification of Eutectoid Steel of 0.90 Per Cent. of Carbon.— Starting at the point h, Fig. 23, which represents this steel when at 2,000°, and hence molten, and following its cooling by passing down this line, the steel begins to freeze as austenite when the temperature has fallen to h^{I} or 1,445°. But freezing, instead of completing itself at this temperature, goes on progressively till the temperature has fallen to h^{II} or 1,205°, so that whereas at all temperatures above h^{I} the steel is wholly molten and at all temperatures below h^{II} it is wholly solid, yet at all temperatures between h^{I} and h^{II} it is a mixture of certain particles which are solid austenite and certain which are still molten.

The quantitative course of this solidification is shown in Diagram II of Fig. 24 as a progressive increase in the quantity of solid austenite as the temperature sinks from 1,445°, or h^{I} on the liquidus to 1,205°, or h^{II} on the solidus.

199. A sketch of the solidification of alloys relatively poor in carbon is given in Diagrams I, II, and III of Fig. 25, §169, I showing the progress of solidification of eutectoid steel of 0.90 per cent. of carbon, II that of a strongly hyper-eutectoid steel of 1.70 per cent. of carbon, and III that of a cast iron 2.40 per cent. of carbon, that is more strongly hypo-eutectic than the 3

[1] See Tiemann, *The Metallographist*, 1901, vol. 4, p. 319.

per cent. carbon one considered in §§176 to 181. In each case there is the progressive solidification of primary austenite, represented by the progressive increase in the primary austenite crosses and decrease in the quantity of residual molten, as the temperature sinks from the liquidus to the solidus; and in the two cast irons there is the solidification of the eutectic at 1,135°. Beyond this, as we pass from right to left the quantity of eutectic decreases from 100 per cent. in Diagram V of the eutectic cast iron, to zero in Diagram II of the 1.70 per cent. carbon steel, as shown in that part of these various diagrams below the 1,135° line *EBC*.

These sketches bring out the close similarity of the course of solidification for all non-eutectiferous alloys of this series, that is of those with less than 1.70 per cent. of carbon, the chief difference being that the range covered by solidification becomes higher and narrower as the carbon content decreases, till chemically pure iron freezes at a single point, about 1,530°C.

200. The incompleteness of the selection in the solidification of hypo-eutectic alloys, both steel and cast iron, should here be noted. To fix our ideas let us consider the case of a steel of 0.90 per cent. of carbon.

The solidification of all these alloys is indeed selective, but this selection differs from the selection of our nitrate solution and of our hyper-eutectic cast iron, in being incomplete where they are complete. That is to say, the austenite which solidifies during this pro-eutectic or primary period, instead of being pure iron, corresponding to the pure nitrate or pure ice of our nitrate-water solutions, or to the pure primary cementite which solidifies during the pro-eutectic period of a hyper-eutectic cast iron, contains carbon actually dissolved in it. Moreover, this dissolved carbon content of the solidifying austenite increases progressively as the temperature falls during the solidification. This percentage should not normally exceed 1.70 per cent., point *E*, and this richness it reaches only in case the conditions cause the solidification to continue down to the eutectic freezing temperature, 1,135°, at which any remaining molten should be of eutectic composition, and hence should now freeze unselectively as eutectic.

201. A Mechanism of this Incomplete Selection.—Let us consider this a little more fully. Turning to Fig. 29 which shows the mechanism of the selective freezing for this 0.9 carbon steel, the carbon content of the depositing layers of austenite is represented by the "solidus" curve *AE*, *i.e.*, the curve which represents the temperature and carbon content of the solidifying layers, under conditions of equilibrium. Thus in our present case, represented by ordinate $h\ h^{\mathrm{I}}h^{\mathrm{II}}$, the solidification begins selectively when, in cooling, the temperature reaches h^{I}, or 1,445°, on the liquidus *AB*, by the deposition of solid austenite of the carbon content represented by *AE* for the same temperature, *i.e.*, the carbon content 0.14 per cent., of the point *q* on *AE* horizontally opposite h^{I}.

This deposition of solid austenite, poorer in carbon than the molten metal out of which it is depositing, of course enriches that molten in carbon. But because the molten is now richer than it was when it deposited that first

layer of carbon content q, the layer which it next deposits is correspondingly richer than q, the carbon content of the layer depositing at any given instant increasing with the carbon content of the molten with which it is in contact. The relation between these two carbon contents is represented by the solidus and the liquidus respectively, so that the carbon content of the depositing layers at any instant is measured by the abscissa of the point on AE horizontally opposite the point on AB, the abscissa of which is the existing carbon content of the layers of molten out of which that deposition is taking place.

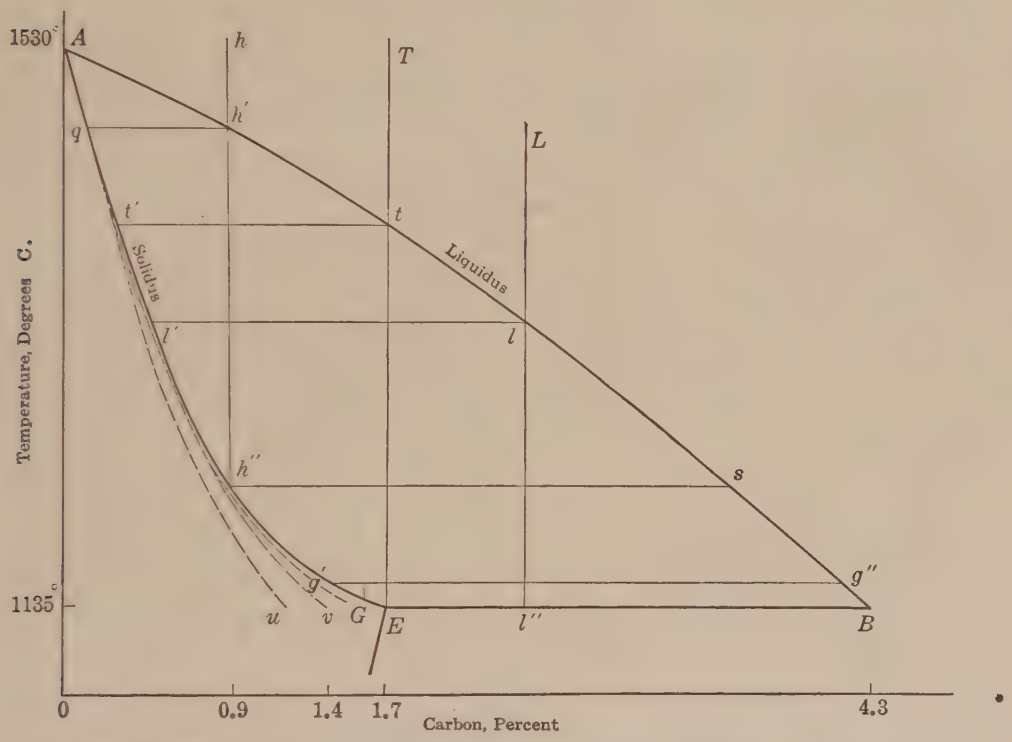


FIG. 29.—Fragment of the carbon-iron diagram showing the mechanism of the selective solidification of primary austenite.

Hence, under conditions of strict equilibrium, at each temperature in the further cooling the carbon content of the molten is represented by the point on the liquidus AB cut by the abscissa of that temperature, and the carbon content of the austenite which deposits from that molten is measured by the point on the solidus AE cut by that same abscissa, so that the molten and the depositing layers respectively may be said to slide down AB and AE, the former from h^{I} to s, the latter from q to h^{II}. Referring the reader for a fuller explanation to Chapter 11 on the phase rule, suffice it here to say that these two curves, the liquidus and the solidus, represent the carbon content, *i.e.*, the concentration, which solid and molten respectively have when they are in true equilibrium with each other at the given temperature, and that, if either had some other carbon content, they would immediately start read-

justing their carbon contents in the direction of equilibrium, either by dissolving part of the solid or by solidifying part of the molten.

Because the solid, after solidification is complete, must needs have the average composition with which the molten metal started, or h, therefore at the end of solidification it is represented by some point on that same ordinate. This point must be on the solidus if equilibrium is complete, or in short the temperature and carbon content at the end of solidification are represented by the intersection of ordinate and solidus, h^{II}.

202. Differentiation in Solidifying and Equalization by Diffusion.—Solidification has thus been seen to be a process of differentiation. The layers which deposit are poorer in carbon than the molten whence they deposit. This must needs be reflected by a corresponding enrichment of the molten mass, and hence by the progressive enrichment of the successively deposited layers, those first deposited being much poorer in carbon than those last to deposit. The axis of the trunk of each pine tree is poorer in carbon than its bark, and the base of each tree, where it roots against the walls of ingot-mould, is poorer than what we may call its tip, its other end. Hence the familiar result known as "segregation," or the enrichment of the central or more accurately the last solidified part of any given ingot or casting, in the three segregating elements, carbon, sulphur, and phosphorus, at the expense of the outer parts which have solidified first. The richest part of the "segregate" usually lies near the bottom of the pear-shaped internal contraction cavity called the "pipe," because in this region the end of solidification occurs.

But as soon as two or more layers thus differing in carbon content have been deposited, diffusion starts in to efface in them the heterogeneousness thus created. If this effacement were complete so as to bring the whole solid to a uniform content, that content must clearly be the initial carbon content of the molten, which we have assumed to be h^{I}. Hence under these conditions the temperature and carbon content of the solid at the end of a solidification slow enough to permit diffusion to efface completely the differentiation which the solidification itself causes, would be represented by the point h^{II}. At the very end of solidification the carbon of the last drop of molten, of temperature and carbon content s, on depositing would diffuse back, and by our present assumption its carbon in excess of h^{II} would spread itself out uniformly over all the previously deposited layers, bringing them to carbon content h^{II}.

This diffusion of carbon is indeed rather rapid, at least in most cases; but the diffusion of phosphorus is extremely slow, and its slowness gives rise to phenomena which give an apparently false suggestion of slowness of carbon diffusion (§§772 to 779).

203. Because Selection is Incomplete, the Alloys Poor in Carbon Contain no Eutectic.—In the case of our nitrate solution and of our hyper-eutectic cast iron, it was the rigidity of the selection, the purity of the pro-eutectic substance, or more exactly its freedom from dissolved nitrate, or water, or iron as the case might be, that implied that every solution or alloy must needs

be eutectiferous, containing enough eutectic to represent the whole of nitrate present in a poor solution, and in a hyper-eutectic cast iron the whole of the iron except that which forms part of the primary cementite.

But in the case of this present hypo-eutectic iron carbon alloy, the eutectoid or 0.90 per cent. carbon steel, we saw that, though the selection which occurs during the solidification does cause a certain progressive concentration of the carbon initially present in the gradually shrinking mass of residual molten, yet this selection and concentration are so incomplete that in spite of them even the last solidifying layers are able to dissolve the whole of their carbon, under the conditions described.

The essential condition which enables this dissolving of the carbon in the solidified austenite to become complete is that, at the end of solidification, the last drop of molten to solidify either shall not contain more carbon than it can retain in solution after it has solidified, or else shall diffuse back into the earlier deposited and poorer layers all carbon in excess of what it can retain in solution. In brief that the carbon content of the last of the solid to solidify shall not exceed permanently its solvent power for carbon.

203A. The Solubility of Carbon in Solid Austenite.—This saturation point, solvent power, or solubility, according to Gutowsky's experiments,[1] is 1.70 per cent. at the temperature 1135°, or the point *E*. As the temperature falls farther this solubility decreases as sketched by the lines *ES*, Fig. 23. As is pointed out in §207, the experimental conditions tend to make the solubility appear less than it really is.

204. The Eutectiferous Range Begins at 1.70 Per Cent. of Carbon under Conditions of Equilibrium.—For simplicity let us still assume that the diffusion of the carbon in the solidified layers completes itself as solidification progresses, so that at each instant they are of uniform carbon content throughout, though of course that uniform carbon content increases as solidification progressess, following the solidus *AE*. And to fix our ideas let us adopt Gutowsky's determination of the solubility of carbon in austenite at the eutectic freezing-point, 1135°, as 1.70 per cent. Under these conditions a steel of 1.70 per cent of carbon, *T* of Fig. 29, will begin to deposit layers of t′ per cent. of carbon when in cooling it reaches t. As cooling proceeds the molten and the solid will slide down the liquidus and solidus respectively, till at 1,135° the former reaches *B* and the latter *E*, or 1.70 per cent. Under our present assumption of perfect diffusion, this will bring the carbon content of the solid to 1.70 per cent., a content which does not exceed its solvent power.

The condition which has permitted this is that, at the moment when the molten has reached carbon content *B*, its quantity has fallen to zero, as is readily seen. At all times during this solidification the molten is richer than the solid in the ratio of the abscissa of the liquidus to that of the solidus. When the solid reaches *E* it has 1.70 per cent. of carbon. But the average of solid and molten jointly must also be 1.70, for that is in the initial carbon content of the molten, *T*. But 1.70 cannot be simultaneously the carbon

[1] *Metallurgie*, 1909, vol. 6, p. 731.

content of the solid, and also the average carbon content of the solid taken jointly with a molten which is richer than that solid, unless the quantity of that molten is zero.

If, on the other hand the initial carbon content of the whole mass is 2.40 per cent., L of Fig. 29, so that during solidification the solid slides down from l^{I} to E while the molten slides from l to B, then when they have reached E and B respectively, there will still be a large quantity of molten left, a quantity which is found to be 26.90 per cent. of the whole by solving the equations, $1.7x + 4.3y = 2.4 \times 100$, and $x + y = 100$, in which x and y = the percentage of solid and molten respectively.

But as the present solid is saturated with carbon, containing 1.70 per cent., it cannot absorb any of the carbon of the present molten. Hence that molten solidifies without reacting on the now existing solid, and, as it was of the eutectic carbon content, 4.30 per cent., it solidifies as the eutectic. The mass is eutectiferous, as is sketched in Fig. 25. Because that which solidified during the earlier part of the solidification from L to l^{II} was primary austenite, whereas that which solidifies at the end of solidification, at l^{II}, is eutectic, that eutectic is found occupying the interstices between the earlier formed pine-tree crystals of that austenite, as in Figs. A, B, and C of Plate 6. Other cases are shown in Figs. J, K, and L of Plate 3, E of Plate 4, A to E inclusive of Plate 12, and A to F of Plate 13.

But by like reasoning any molten which contains more than 1.70 per cent. of carbon will in solidifying give rise to more or less eutectic, the quantity of eutectic being calculable by the formula just given, and varying from 0 per cent. of the whole when the alloy as a whole contains 1.70 per cent. of carbon, to 100 per cent. of the whole when the alloy contains 4.30 per cent. of carbon. Remember that we are still assuming that diffusion in the solid effaces completely the differentiation which arises during solidification.

205. The eutectiferous range is shown micrographically in Figs. I to O, Plate 3. Note that Figure M shows eutectic only, save for a very small patch of primary cementite at the lower left-hand corner; that Fig. I of a very high-carbon steel, shows primary austenite only, here metamorphosed during slow cooling into primaustenoid, a mixture of pearlite and pro-eutectoid cementite; and that the intermediate figures show a progressively increasing quantity of austenitic space or primaustenoid and a decreasing quantity of eutectic. These changes of course come to an end when the carbon content falls to 1.70 and the quantity of eutectic to zero, and its further decrease only changes the carbon content of the austenite, which forms the whole of the mass.

Fig. I,2 of Plate 3 shows a structure intermediate between those of I and J of Plate 3, that of a eutectiferous or extra high-carbon steel. Here the eutectic is represented by the white masses parallel to the direction of rolling, some of which have the honeycombed structure of the eutectic.

Passing in the opposite direction, Figs. N and O show hyper-eutectic cast irons, with their great white masses of primary cementite.

The expediency of making the end of the eutectiferous range serve as a boundary between steel and cast iron is discussed in §49, Chapter 4. The present end, 1.70 per cent. of carbon, certainly does not recommend itself as strongly for this purpose as the end which we formerly accepted, 2.20 per cent., because industrial alloys with between 1.70 and 2 per cent. are both malleable enough and ductile enough to be associated more readily with the steels than with the cast irons (§102).

206. Industrial Cast Irons in General and very High Carbon Steels are Eutectiferous.—We have seen (§192) that the complete selection in the solidification of primary cementite makes hyper-eutectic cast irons eutectiferous, and (§170) that eutectic cast irons solidify wholly as eutectic. The only remaining class of industrial pure carbon cast iron is hypo-eutectic, and all such hypo-eutectic cast irons contain more than 1.70 per cent. of carbon, and hence from the preceding section are eutectiferous. This eutectic may of course become metamorphosed by graphitization.

This assertion is limited to industrial cast irons, which in fact are never strongly hyper-eutectic, because there is enough doubt about the constitution of strongly hyper-eutectic ones to prevent any positive assertions about it. In particular the constitution of those containing more than 6.67 per cent. of carbon, that is those richer than cementite, is in dispute.

207. Imperfect Diffusion Widens the Eutectiferous Range.—Returning to our steel of 1.70 per cent. of carbon considered in §199, let us remember that the reason why no eutectic formed was that diffusion had by assumption equalized completely the carbon content of the solid part, and had continued this equalization up to the moment of the deposition of the last particle, when the molten had been enriched up to the eutectic carbon content of B or 4.30 per cent., and that it was because of this complete equalization that the carbon content of the molten could not reach 4.30 per cent. till the quantity of molten had fallen to zero.

But it is only under these conditions of complete equilibrium, including this assumed complete equalization of the carbon content of the solid part at each instant during solidification, that the successive temperatures and carbon contents of the solid are represented accurately by $t'E$ of Fig. 29. Clearly such a completion of equalization can be approached only asymptotically. With any usual rate of cooling and solidification the path of the solid layers must be represented not by any single line, but by a fan-shaped area such as $t'uE$. Here $t'u$ may represent the changes in carbon content of the first deposited layer with falling temperature, its progressive enrichment by diffusion from the later deposited and richer layers. The lines between $t'u$ and $t'h''E$ represent later and later deposited layers, and E represents the layer deposited at the instant of the arrival at 1,135°, simultaneously with the arrival of the molten at carbon content B.

Fixing our attention on this moment, the average carbon content of the solid layers deposited up to this point will clearly be somewhere between u and E. To fix our ideas let us take it as $v = 1.4$ per cent., though the exact

point assumed is immaterial for our present purpose. The average carbon content of molten and solid jointly being 1.70, by solving the equations

$$1.4x + 4.3y = 1.7 \times 100, \text{ and}$$
$$x + y = 100,$$

we find that there is at this moment 10.30 per cent. of molten remaining.

But because the last of the layers thus far deposited, the layer with which the molten is now in contact, contains 1.70 per cent. of carbon and is therefore saturated, it cannot absorb any further carbon from the molten, which is inert toward it, and hence solidifies unselectively as eutectic. Hence the mass at the end of the solidification consists of 89.70 per cent. of primary austenite and 10.30 per cent. of eutectic. In short, the imperfection of diffusion in the solid makes this steel eutectiferous, though under conditions of equilibrium, including complete equalization of the carbon content of the solid, it would not be eutectiferous.

It is true that during the solidification of this quantity of eutectic there would be some further diffusion of carbon content out from what were the shore layers of solid at the moment when the molten reached the eutectic carbon content of 4.30, permitting those layers to absorb a little carbon from the eutectic, and this diffusion would lessen the quantity of eutectic proportionally. Yet it is clear that unless this equalization by diffusion becomes complete, there must be some quantity of eutectic which those layers will be unable to absorb, because they cannot absorb more than 1.70 per cent., and hence that some eutectic will persist.

By like reasoning we see that, even with much smaller carbon contents, if the equalization of the carbon content of the solid is sufficiently imperfect to bring their shore layer to 1.70 per cent. of carbon before solidification is complete, the molten then remaining and containing 4.30 per cent. of carbon cannot give up any of its carbon to these saturated layers, but will be inert toward them, and hence will solidify as eutectic with 4.30 per cent. of carbon.

208. This Widening Applies to Other Methods of Determining the Width of the Eutectic Range.—Because of the difficulty of bringing about perfect diffusion in the solidifying mass, the true equilibrium limit of the eutectiferous range is hard to determine by direct observation, which tends to make it wider than it would be if equilibrium were reached. Even the alternative method of "eutectic times" is affected by this same source of error. This consists in solidifying a series of alloys differing in carbon content, under identical outward conditions, and recording accurately the progress of their cooling. On reaching the eutectic freezing-point, 1,135°, the eutectic itself freezes, and the heat which its solidification evolves arrests the cooling, holding the temperature constant at or very near 1,135°. The length of time that the temperature thus remains stationary is very nearly proportional to the quantity of eutectic which forms, so that by extrapolation we can find the carbon content at which this time would fall to zero and the quantity of eutectic would become zero. Or, to speak more generally, plotting against

the carbon content the heat evolution of eutectic solidification implied by the distortion of the cooling curve there, and extrapolating, we reach a carbon content at which this heat evolution falls to zero, and with it the quantity of eutectic formed. But this method too is evidently subject to this same error from the formation of undue eutectic.

209. The Widening of the Eutectiferous Range shown Micrographically. —Because it is the imperfection of diffusion that widens the eutectiferous range, and because rapid solidification and cooling by shortening the time available for diffusion tend to make it imperfect, we naturally look for evidence of this widening in alloys which have solidified rapidly. Such cases are shown in Figs. A, B, and E of Plate 7. These show steels which, under conditions of equilibrium, would be free from eutectic, because their carbon content is less than the 1.70 per cent. which forms the normal boundary of the eutectiferous range. Yet they all show the presence of eutectic, caused by their very rapid solidification. Of these Figs. A and B show part of a 1-in. test ingot of steel of about 1.08 per cent. of carbon quenched in water,[1] and Fig. E part of a mass of steel of unknown carbon content, solidified very rapidly by pouring it on thick masses of cold iron.

The position of the eutectic is to be noted particularly. We shall see that, in solidifying metals in general, the region of last solidification is at or somewhat below the bottom of the internal cavity left by the ebb of the molten, an ebb caused by the different rates of cooling and hence of contraction of the various concentric layers. This cavity is called the "pipe." In the cases shown in Figs. A, B, and E, Plate 7, the eutectic lies close to the pipe, in accordance with the teaching of §207 that the enrichment is greatest in the layers last to solidify.

The region shown in Fig. B, Plate 7, contains about equal parts of eutectic and of primaustenoid, and hence its carbon content is about $(1.7 + 4.3) \div 2 = 3$ per cent.; so that lack of time for diffusion to efface the concentration which arose in solidification has here left the last solidifying parts with nearly thrice their normal carbon content of 1.08 per cent.

Just as hastening the solidification widens the eutectiferous range and thus may give rise to the presence of eutectic in a steel which, if solidified more slowly, would have had none, so after a steel has solidified at a normal rate without forming any eutectic, if we re-melt it even slightly and then solidify it more rapidly than before, we can form some eutectic within it. But the evidence showing this can be studied to better advantage in §264, after familiarizing ourselves with the transformation.

[1] The Author, *Trans. Amer. Inst. Mining Eng.*, 1909, vol. 40, p. 644, Figs. 3 and 5.

CHAPTER 9

INTRODUCTION TO THE CARBON-IRON DIAGRAM (CONTINUED)

The Transformation in Steel

210. The Transformation Range in General.—By "the transformations" we mean those specific ones which both steel and cast iron undergo, not only after they have solidified completely but in the case of most steels at or about a red heat, or at least at temperatures very far below that of complete solidification. They are the transformations by which solid steel in cooling slowly passes from the non-magnetic, high-temperature, austenite state to the low-temperature state of pearlite plus ferrite or cementite, the transformations by obstructing which rapid cooling hardens steel. As we have seen, the temperature at which they occur is represented by the "transformation curves" *GOSE, PSK,* Figs. 23 and 24.

Note first that the shape of the group of lines which forms the transformation curves, Figs. 23 and 24, is of the same family with that of the solidification curve *ABD, EBC* of these same diagrams, and with the freezing-point curve *ABC, aBc* of our nitrate-water series, Fig. 20. Second, that the transformations themselves are strictly comparable with the freezing of the nitrate series, and almost as strictly comparable with the solidification of our molten alloys. Thus the carbon iron diagram really consists of two closely comparable sets of curves, the upper one *ABD, EBC* representing solidification, and the lower one, *GOSE, PSK,* representing closely comparable transformations within the red-hot solid metal. Indeed it is chiefly as a preparation for the study of the transformation curves that these other curves have been studied so closely (§§168 *et seq*).

In particular, the transformations are (1) *selective;* (2) beginning when the cooling reaches the *liquidoid GOSE;* (3) with the separation of the excess or *pro-eutectoid* substances; (4) at such a rate that the residual untransformed (solid) austenite mother metal slides down the liquidoid, its carbon content and temperature being at all times represented by abscissa and ordinate of these curves; till (5) it reaches the eutectoid carbon content and transformation temperature, quite as the residual molten part of our solidifying alloys thus slid down its liquidus, till its carbon content and temperature simultaneously reached that of the eutectic; when (6) its transforms into the composite eutectoid *pearlite* (§§71, 74), quite as the eutectic molten solidified into the composite eutectic *ledeburite.*

211. Hardenite.—Austenite of this eutectoid carbon content, 0.90 per cent. is often called hardenite.

212. A1, A2, A3, etc., the boundaries of the transformation range, are named as follows: The lower boundary *PSK* is called *A*1, or in falling temperature the *recalescence.* The hypo-eutectoid part of the upper boundary

GOS is called *A*3. *The line* or zone at *MO*, at which the change from beta to alpha ferrite in cooling down and the reverse change in heating up chiefly occur, is called *A*2 (§228).

When the carbon-content is between about 0.48 and 0.90 per cent., the ferrite which forms in the transformation seems to pass at once to the alpha state, so that here the lines *A*3 and *A*2 seem to coincide; hence this part, *OS*, of the upper boundary is called *A*2-3.

213. In hyper-eutectoid steel and cast iron the line *SK* may be called *A*3-2-1,on the ground that at this line the transformations identified with *A*3, *A*2, and *A*1 respectively occur here also, in the transformation of the residual austenite, here brought to the eutectoid carbon content 0.90, into pearlite. For the pearlitic ferrite as it is here generated changes from the gamma state in which its iron existed above *SK* to that of alpha ferrite, *i.e.* undergoes the gamma-alpha change, represented by $A3 + A2$; and the change from eutectoid austenite into pearlite itself is represented *A*1. By like reasoning the point *S* itself may be called *A*3-2-1.

214. The Ae or Equilibrium Points.—For each of these points, *A*1, *A*2, etc., there is an equilibrium temperature at which the change which it represents is due, though this change may not actually set in till, in heating up, the temperature has passed above this equilibrium point, and in cooling down it does not actually occur till the temperature has passed below this point. This equilibrium temperature may be called *Ae*, so that *Ae*1 represents the temperature at which the change *A*1 is theoretically due, the exact temperature at which it would occur were there no lag whatsoever. *Ae*1 *Ae*2, and *Ae*3 thus represent respectively the equilibrium temperatures for the recalescence in cooling and the absorption point in heating, the great loss and the great recovery of the magnetic properties in heating and in cooling, and the end of the re-absorption of ferrite in heating and the beginning of its precipitation in cooling. The actual temperatures of these several phenomena approach these equilibrium temperatures more and more closely as the lag is lessened more and more, but they can never pass them, and we may doubt whether they can ever attain them exactly. There is indeed reason to suspect the presence of a labile range, which might serve as a positive bar to making the transformations in cooling occur at the same temperature as in heating up.

215. The Ac and the Ar Points.—The temperatures at which these various changes actually occur in any given case, as distinguished from the *Ae* points at which they are due, are called the *Ac* and *Ar* points to distinguish them from the *Ae* points at which they are due in equilibrium.

The Ac points are the actual change-temperatures when the steel is heating up.

The Ar points are the actual change-temperatures when it is cooling down, "*c*" referring to "*chauffant*" and "*r*" to "*refroidissant.*"

216. Double Meaning of these Symbols.—These various symbols, *Ac*, *A*3, etc., refer either to the actual temperature at which a given transformation

occurs, or to the transformation itself. The context should and generally does indicate which sense is intended.

"The recalescence," literally "the reheating," was first used specifically as meaning the change *Ar*1 from austenite to pearlite, which under favorable conditions causes a visible rise of temperature, a brightening of the now dull red steel. I recommend restricting it to this specific sense, because it is already identified with this change, because nothing is to be gained by giving it an additional and generic sense for which "transformation" already suffices, and because the other transformations are not strictly recalescences, in that they cause no rise of temperature.

217. The Absorption Point.—The temperature of the reverse change, *Ac*1, is sometimes called the "absorption temperature" or "absorption point," because of the important absorption of heat which occurs in heating up past it. This change clearly represents the acquisition by hypo-eutectoid steels of part of the hardening power, and by hyper-eutectoid ones of the whole of it.

218. The parallelism of the eutectoid, pearlite, to the eutectic, ledeburite, already set forth in §86, is here confirmed as regards the constancy of the genetic temperature, irrespective of the carbon content of the alloy. Recalling their parallelism, (A), in that each is of constant composition and genetic temperature irrespective of the carbon content of the alloy; (B), in that neither has simple atomic proportions; and (C) in that each is a mechanical mixture of those two substances of which one or the other accompanies it in alloys of higher or lower carbon content than its own, we here add that each is the alloy of lowest genetic temperature of the whole series. For pearlite has the lowest formation temperature, about 725°, *S*, or *Ar*3-2-1 of Fig. 23, and ledeburite has the lowest solidification temperature, 1,135°, or *B* of that figure. More strictly the genesis of ledeburite occurs at the temperature at which the solidification of all other pure carbon eutectiferous alloys ends; and the temperature at which the whole genesis of pearlite occurs, in equilibrium, is that of the end of the transformation of all pure carbon steels of whatever carbon content.

Thus the eutectoid is to the transformation exactly what the eutectic is to solidification. Indeed eutectoids were formerly called "eutectics," but this led to such confusion that I reluctantly proposed the term "eutectoid"[1] to denote the alloy of lowest transformation point, and thus to distinguish it from the eutectic or alloy of lowest melting-point.

If we substitute "transform" for "solidify," "eutectoid" for "eutectic," and "pro-eutectoid" for "pro-eutectic or primary," we can describe the phenomena represented by these two sets of curves, those of solidification *ABD*, *AEBC*, and those of transformation *GOSE*, *PSK*, in otherwise identical terms.

Though this constancy of the transformation temperature of the eutectoid is rigidly true under conditions of strict equilibrium, as follows from the

[1] *The Metallographist*, 1903, vol. 6, p. 249. Desch, "Metallography," Longmans, London, 1913, Second Edition, p. 59.

phase rule, yet because the undercooling or lag (§222) is greater in low- than in high-carbon steel, the observed position of this transformation rises slightly with the carbon content.

218A. Points of Difference between Pearlite and Ledeburite.—There is indeed the difference that, whereas no matter how rapidly or slowly molten iron solidifies, the ledeburite takes on its typical honeycomb structure, pearlite does not take on its typical structure if the cooling is too fast, and it may not retain it if the cooling is too slow. Indeed, pearlite does not come into existence at all if the cooling is rapid. But these contrasts are necessary results of the difference in the attendant conditions.

A rapid cooling prevents pearlite from coming into existence because the transformation, of which it is a product, is capable of suppression by rapid cooling; but a rapid cooling cannot prevent ledeburite from coming into existence, because solidification is not suppressible.

A cooling which is barely slow enough to permit the transformation yields, not typical lamellar pearlite, but pearlite in the state of an emulsion irresoluble microscopically, called sorbite, whereas ledeburite is honeycombed no matter how fast it solidifies, as is seen in Figs. O of Plate 3 and C of Plate 6, of white cast iron poured into water while molten. The reason for this contrast is that ledeburite, as it forms, has the great mobility of the molten state out of which it springs, whereas pearlite has only the very moderate mobility of the relatively rigid and completely solid state at 725°, about 400° below the temperature of complete solidification at which ledeburite forms.

It is because the transformation differs from solidification both in being suppressible and in occurring in a rigid instead of a molten mass, that a further structural difference between steel and cast iron arises. A rapid cooling suppresses the birth and coalescence of the pro-eutectoid element of steel, so that a very rapidly cooled steel has no free pro-eutectoid element, and one cooled only barely slowly enough to permit the transformation has no visible pro-eutectoid element, which in the former case forms part of the martensite and in the latter part of the sorbite. But even the most rapid cooling cannot prevent the idiomorphic arrangement of the primary element of cast iron. Thus the rapidly cooled steel of Fig. A of Plate 2 has no pro-eutectoid element but only martensite and austenite. In the rapidly cooled steel grains of Fig. B of Plate 10 there is no pro-eutectoid cementite, but only martensite; whereas the eutectic masses between those grains, which are equivalent locally to cast iron, show their conglomerate structure clearly, their eutectic austenite and eutectic cementite. Again, the equally rapidly cooled cast irons of Fig. O of Plate 3, and Fig. C of Plate 6, both poured while molten into water, have not only the typical honeycombed structure of the eutectic but also the idiomorphic shape of the primary element, primary cementite in the former and primary austenite in the latter.

Finally a very slow cooling breaks pearlite up, or divorces it, whereas the structure of ledeburite is firmly fixed, and can be broken up only by a very long exposure to a temperature far above that at which pearlite divorces.

The reason for this is that the plastic ferrite forms nearly nine-tenths of the substance of pearlite, and thus dominates it. The relatively rigid cementite is in such a small minority that it has little power to resist the surface-tension efforts of the ferrite to slip from the embrace of the pearlitic cementite. That cementite does not isolate the ferrite masses from each other. But in the ledeburite the cementite is in such great excess, and it so isolates the austenite in the form of separate comb fillings, as to obstruct their attempts to ball up by surface tension. Hence the structural instability of pearlite and stability of ledeburite.

The movements which do occur in the cementite masses of pearlite may indeed be due in large part to re-solution and re-precipitation.

219. The Most Important Lines.—For hypo-eutectoid steels the most important line is *A*3 or *GOS*, and for hyper-eutectoid ones the most important is *A*3-2-1, the recalescence, or *PSK*.

*A*3 is the temperature above which hypo-eutectoid steel must be heated in order that it may be either fully hardened by rapid cooling, or grain refined for the purpose of improving its structure. *PSK* is the temperature above which hyper-eutectoid steel must be heated in order to be hardened properly by rapid cooling.

220. General Diagram Illustrating the Microstructural Changes of the Transformation.—Fig. 30, after Stead,[1] shows in a general way the microstructural changes during the transformation, by means of the figured part within the lines *GOS* or *A*3 and *SE* above, and *PSK* or *A*1 below. Its lettering corresponds to that of Figs. 23 and 24.

These changes are of two kinds, those in the white pro-eutectoid network, and those in the matrix or meshes of pearlite which that network encloses. The network begins to be generated when the temperature cools past *A*3 or *SE*, the upper limit of the transformation range, and its walls gradually thicken in cooling down through that range; and it is gradually re-absorbed by the matrix in heating up through that range.

The matrix consists of austenite at all temperatures above *A*1, here shown in black, and of pearlite at all temperatures below it, here shown in stripes, with an abrupt change, *the recalescence*, from austenite to pearlite on cooling past *A*1, and an abrupt change, *the absorption*, back from pearlite to austenite, on heating past *A*1.

The network consists of ferrite or nearly pure iron in steels with less than 0.90 per cent. of carbon, called hypo-eutectoid steels, and of cementite, Fe_3C, in steels with more than 0.90 per cent. of carbon called hyper-eutectoid. The black austenite as we have already seen is apparently a solid solution of cementite in gamma iron (§95), while pearlite is the eutectoid conglomerate of sheets of pearlitic cementite, Fe_3C, with pearlitic ferrite, in the

[1] "The Practical Use of the Iron-carbon Equilibrium Diagram, etc.," *Proc. Northeast Coast Institution of Engineers and Ship-builders*, 1912-13, p. 29. See also *Sixth Report to the Alloys Research Committee, Instit. Mechan. Engineers*, 1904, Fig. 163 on Plate 40; and *Journ. West of Scotland Iron and Steel Institute*, 1912, vol. 19, Fig. 34, following p. 204.

ratio of about 1 of the former to 6.4 of the latter (§§71, 74). We will consider these changes more fully in the sections immediately following, but note here again that the upper limit of this transformation range descends from about 920° to 725°, and again rises to 1,135°, as the carbon content rises from 0 to 0.90 per cent., and then to 1.70 per cent.

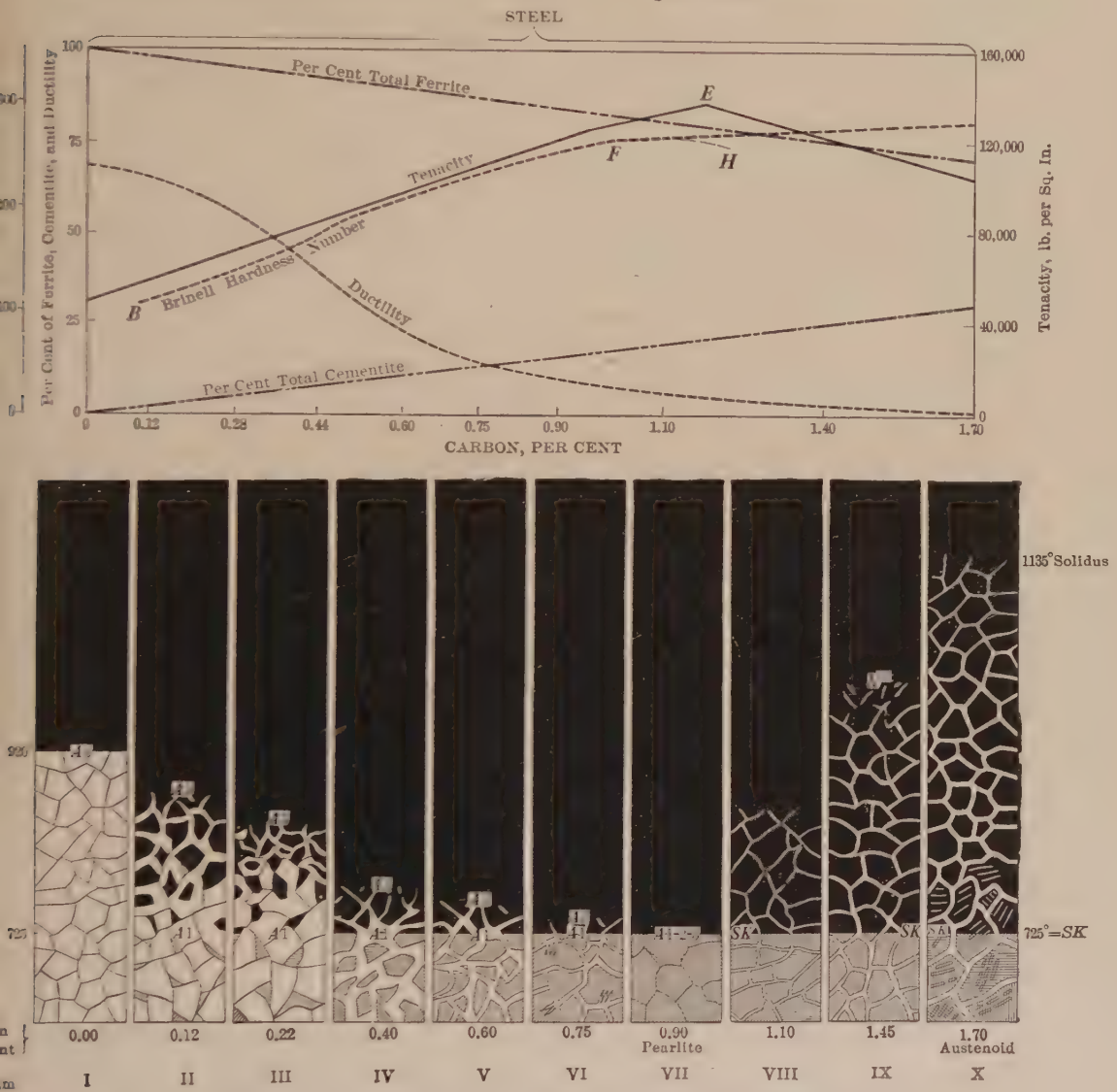


FIG. 30.—Changes in the structure of steel during the transformation, *A*3 to *A*1.

The teachings of this diagram will be reinforced by those of many detailed micrographs in what follows.

Just as we studied the solidification of eutectic, hypo-eutectic, and hyper-eutectic cast iron in §§170 to 197, so let us now study the transformation of eutectoid, hypo-eutectoid, and hyper-eutectoid steel in §§221 to 263, and that of cast iron in §§266 to 277.

221. The Transformation of Eutectoid Steel of 0.90 Per Cent. of Carbon in Cooling.—Taking up the cooling of our 0.90 per cent. carbon steel at the point at which we left it in §198, when it had solidified completely and reached h^{II} on the ordinate of Figs. 23 and 24, the further cooling goes on with no change which we need consider till the solid metal cools to S, as is indicated in Figs. 23 and 24 by the absence of lines crossing $h^{I}h^{II}$ and in Fig. 30 by the unbroken continuity of the black austenite from h^{II} to S. But, as the steel starts to cool past S, the austenite of which it has hitherto consisted is suddenly transformed into pearlite, with such an evolution of latent heat that the now barely red-hot metal brightens strikingly, and is said to "recalesce."

This transformation of the eutectoid at one temperature (see §§86, 210, and 218) into pearlite is strictly comparable with the solidification of the eutectic nitrate solution, and of the eutectic or 4.30 per cent. carbon cast iron into ledeburite, each of which also occurred at one temperature. It is shown quantitatively at the eutectoid line 725° of Diagram II, Fig. 24.

It is of the first importance. The hardening of steel by rapid cooling consists in cooling past S so fast as to impede the change from the austenite state which exists above S to the pearlite state into which the steel tends to pass below S, with the result that this change is arrested in transit, catching the steel in the state of martensite, intermediate between austenite and pearlite, and harder than either. Martensite is the familiar condition of glass-hardened steel (§97). The recalescence represents the complete change from austenite past martensite to pearlite, a change which finds time to complete itself during slow cooling. Once steel has undergone this recalescence in cooling slowly to ever so little below S, it cannot be hardened even by the most rapid quenching. *The recalescence thus represents the loss of the hardening power.*

In further cooling no further change which we need consider occurs, and the metal when cold consists of pearlite.

221A. Transformation of Eutectoid Steel in Heating up.—If the steel is now heated up again, it goes through these changes in the reverse order, the pearlite changing to austenite when it has risen as high as S, the austenite beginning to melt when it reaches h^{II}, and more and more of it melting till at h^{I} the last of it has melted.

222. Lag.—For simplicity I have thus far spoken as if these transformations really occurred as it were strictly on schedule time, *i.e.*, at exactly the temperature at which they ought to occur, and would occur were equilibrium truly complied with. In fact, like most transformations within solid metals, they lag, at least in cooling. Thus, the temperature at which the recalescence or change from austenite into pearlite is due to occur is about 725°; but under the usual conditions of cooling there is a lag of about 35°, so it does not actually set in till the temperature has sunk to about 690°. For several reasons there is less lag in heating up, and possibly there may be none.

It is by increasing the lag that rapid cooling causes the hardening of steel. The lag is then so great that, before the change to pearlite has time to occur, the temperature has fallen so low, and the molecular state has hence become so rigid, that this change is obstructed and in large part prevented.

As implied in §221, the heat evolved by the transformation raises the temperature of the mass toward or possibly even to that at which the transformation would occur were there no lag, but strict equilibrium instead. Hence, though there is one temperature, about 725°, at which this transformation is due, in fact there is a moderate range of temperature in any part of which it may actually start, and it may actually cover a moderate range.

222A. The quantitative course of the transformation of this 0.90 carbon steel, shown graphically in Diagram II of Fig. 24, §169, by the line *SS*, is the simplest possible, the change of the whole of the black austenite into the striped pearlite at a temperature which is here shown constant, as in the condensation of steam into water or the freezing of water into ice, ignoring the temperature perturbations caused by lag as just described.

223. The microstructural changes during the transformation of eutectoid steel are sketched in Diagram VII of Fig. 30 and shown by means of actual micrographs in Figs. F, G, and H of Plate 8. Note in Fig. 30 the abrupt change at *A*1–2–3 from the black austenite above to the striped pearlite below.

Figs. F, G, and H of Plate 8 represent three points in the length of a bar of steel of 0.92 per cent. of carbon, which was differentially heated[1] so that one end was at about 1,000° and the other end below 600°, with a flat thermal gradient from one end to the other. It was then quenched in water to preserve as far as might be the structure thus set up in its various parts. Those[2] parts which at the moment of quenching were above *Ac*1–2–3 were then in the condition of austenite; those which were below that temperature were in the condition of pearlite.

Fig. F shows the structure of a part which, at the moment of quenching, was far above the transformation temperature, *Ar*3–2–1. Fig. G represents a point which was somewhat cooler though still above *Ar*3–2–1, and H shows one which was below *Ar*3–2–1. Note the coarser martensite masses

[1] Before this differential heating the steel had been very highly heated in order to coarsen the grains of austenite, and then cooled slowly. The pearlite into which it transformed in this slow cooling naturally retained the coarseness of structure which had been given to the austenite, for the purpose of setting up this coarse and hence readily recognized structure.

[2] As far as I know, this method of preserving by quenching, for study in the cold, the conditions existing at high temperature, was used first in the preparation of my 1893 paper, "The Heat Treatment of Steel" (*Trans. Amer. Inst. Mining Eng.*, 1893, vol. 23, p. 466). My paper, "The Hardening of Steel" (*Journ. Iron and Steel Inst.*, 1895, vol. 48, No. II, p. 258), and the paper by Professor Sauveur and myself, "Further Notes on the Hardening of Steel" (*idem*, 1896, vol. 49, No. I, p. 170), were based wholly on it. In this last paper the method was used for studying the high-temperature microstructure, preserved as far as might be by the quenching. But all such advances are the fruit of the time and not of the man. This method was a natural outcome of my wish to test Osmond's theories, and was indeed foreshadowed by his experiments ("Transformations du fer et du carbone," 1888). See Desch, "Metallography," 1913, Second Edition, footnote 6, p. 7.

of Fig. F, the finer ones of Fig. G, and the complete change of structure between G and H. It is this latter change that represents the transformation from pearlite below to austenite above the transformation range, or in this case the transformation temperature, *Ac*1–2–3, the change which is the reverse of the recalescence, and as we have seen is sometimes called "the absorption," and represents the acquisition of the hardening power. Thus the steel of Figs. F and G, because it is now martensite, has become hardened by the rapid cooling, whereas the steel of Fig. H, because it is now pearlite and hence was pearlite at the moment of quenching, has not been hardened by that same cooling.

That it is not austenite but martensite that Figs. F and G show, reflects the fact that in spite of the rapidity of cooling, and the retarding action of the relatively small carbon content, 0.92 per cent., the transformation proceeded from the initial austenite state as far as the intermediate martensite state. Had there been as much as 1.50 per cent. of carbon, this greater quantity of this retarding element would have prevented the transformation from going so far, and the metal would have been a mixture of part austenite untransformed, and part martensite resulting from the incomplete transformation of the rest of the austenite, as is seen in Fig. A, Plate 2. Had there been in addition as much as 2 per cent. of manganese, and had the cooling been extremely rapid, the whole of the austenite might have been preserved, without any transformation into martensite, as seen in Fig. B, Plate 39. (See §§95 to 100, and 240.)

224. Solidification and Transformation of Hypo-eutectoid Steel of 0.40 Per Cent. of Carbon, Ordinate j-j^{VI} of Figs. 23 and 24.—Starting as before at 2,000°, and following the cooling, the molten steel when it has cooled down to j^{I} begins ejecting frozen particles of austenite poorer in carbon than itself, and continues this ejection, and thereby enriches itself in carbon, till it becomes completely frozen when it has further cooled to j^{II}. It reaches the liquidoid or line *A*3 at j^{III}

The quantitative course of this solidification is measured in Diagram I of Fig. 24, §169, by the abscissæ of the curve $j^{I}j^{II}$, solidification beginning at j^{I} and completing itself at j^{II}.

225. Transformations of Hypo-eutectoid Steel, (A) in Cooling.—When this 0.40 carbon steel in cooling reaches the line *A*3 at j^{III} in Fig. 23, or 805°, the transformation begins. It consists of two parts, the *selective* part, and the *eutectoid* part, called also the austenite-pearlite inversion, or recalescence.

226. The Selective Stage of the Transformation.—Just as the selective solidification of our hypo-eutectic nitrate solution consisted of the progressive freezing out of the pro-eutectic or primary ice, and as that of our hypo-eutectic cast iron of 3 per cent. of carbon consisted of the progressive solidification of pro-eutectic or primary austenite, so the selective transformation of this hypo-eutectoid steel consists of the progressive birth within the austenite of particles of the pro-eutectoid substance, pure iron, or "ferrite" *i.e.*, the substance present in excess of the eutectoid composi-

tion of carbon 0.90, iron 99.10 per cent. The way in which these little particles of ferrite are formed, shown in Figs. O, P and Q of Plate 8, will be described more fully in §233.

This process continues from j^{III} to j^{V} in Fig. 23, each grain of mother austenite ejecting more and more ferrite as the temperature sinks, and of course growing proportionally richer in carbon through this rejection of carbonless ferrite, at such a rate that at each successive temperature its carbon content is measured by the abscissa of the line *GOS* of Fig. 23.

That is to say, at each temperature between j^{III} and j^{V} the carbon content of the residual austenite is represented by the abscissa of *GOS* at that temperature. Thus when the temperature has reached j^{V}, the residual austenite has the composition S, which is that of pearlite, 0.90 per cent. of carbon.

227. The Eutectoid Stage.—Then follows the eutectoid stage of the transformation. On now cooling past point j^{V} that residual austenite splits up into pearlite, exactly as the austenite of our 0.9 per cent. steel did at this same temperature, and with like but feebler recalescence, feebler because the recalescing pearlite is diluted by the pro-eutectoid ferrite.

The steel from this point down cools without further change which we need notice, and when completely cold is found to be the common mixture of ferrite and pearlite of which our commercial structural steels consist.

228. Beta Iron.—The passage down through the transformation range has a further complication. There is much to suggest that the ferrite generated while the temperature is above the line *MO*, or perhaps zone *MO*, is beta ferrite; that that generated after the temperature has sunk below this line or zone is alpha ferrite; and that, as this change occurs in the nature of the ferrite which is generated, the same change occurs in the ferrite already generated in cooling from *GO* to *MO*.

The evidence as to the nature of beta iron is extremely difficult to interpret, even by those who seek the truth rather than the victory of an espoused cause. Osmond, who first inferred its existence, would jestingly call it "Decipium" (§§237 to 254).

229. Transformations of Hypo-eutectoid Steel, (B) in Heating.—If this same steel, which has undergone in cooling the transformations described in §§226 and 227 is next reheated, it passes through these same changes in the reverse order. That is to say, its pearlite changes back into austenite abruptly when the temperature rises past 725°, so that as before the change in heating up occurs some 25° higher than in cooling down; and as the temperature rises from j^{V} to j^{III} this austenite progressively reabsorbs the ferrite which it had rejected in the prior cooling till, on rising past j^{III} or a slightly higher point, this reabsorption and the return to the state of austenite have become complete. During further heating this austenite persists, its grain size naturally increasing more and more with the increasing mobility caused by the higher temperature till, on passing j^{II}, certain particles of the steel melt, and with further rise of temperature

more and more particles melt, till at j^{I} all have melted, and our cycle is complete.

230. To sum this up, the stages passed through are:

In region 1 the steel is molten.

In region 2 it consists of a mushy mechanical mixture of particles of solid austenite and other particles of still molten steel.

In region 4 it consists of solid austenite alone.

In regions 6 and 7 it consists of a mechanical mixture of solid grains or kernels of mother austenite together with solid particles of ferrite which those grains have rejected more or less fully to their exterior in the form of cell walls.

In region 8*A* it consists of a mechanical mixture of the ferrite thus ejected, with pearlite resulting from the transformation of the residual austenite at *Ar*1.

231. The quantitative course of this transformation of 0.40 carbon steel is shown at $j^{III}j^{V}$ of Diagram I, Fig. 24, as the progressive replacement of part of the initial black austenite by the white ferrite.

Note that in cooling from j^{III} or *A*3 to j^{V} or *A*1 the austenite gives birth at a slightly varying rate to the pro-eutectoid ferrite, shown as the white left-hand column, the total quantity of that ferrite being measured by the abscissa of j^{V}. On cooling past *A*1 the whole of the austenite then remaining transforms at constant temperature into pearlite, represented by the right-hand striped column, the progressive birth of ferrite in cooling from *A*3 to *A*1 contrasting with this constant temperature of the transformation which follows it at *A*1.

232. A sketch of the microstructural changes in the transformation of steel of 0.40 per cent. of carbon is given in the space between *A*3 and *A*1 in Diagram IV of Fig. 30, which represents this steel. Note the gradual thickening from *A*3 to *A*1 of the white network representing the ferrite, enclosing black meshes which are of austenite. Note that at *A*1 this thickening of the meshes ceases, and that the black austenite is completely replaced below this line by the striped pearlite.

233. Micrographs Illustrating this Transformation of 0.40 Carbon Steel in Cooling.—The way in which the precipitation or expulsion of the pro-eutectoid ferrite begins, with the appearance of minute, bright, yellowish white, metallic looking spots of ferrite is shown in Figs. O, P, and Q of Plate 8. The specimens here shown were cooled slowly to various points in the transformation range, held there 30 and in some cases 60 minutes, and then quenched so as to preserve the structure which had developed during this sojourn. Immediately before quenching the mass consisted of a matrix of austenite, in which these minute particles of ferrite had precipitated. In the quenching this austenite transformed into martensite, which is here represented by the black ground mass through which these light areas are scattered. The contrast between the dark of the ground mass and the light of the ferrite has been strengthened by changing the martensite into the black-etching

troostite, by means of a gentle reheating to 300° or 400° which does not sensibly affect the ferrite itself.[1]

The further progress of precipitation is shown by Figs. A to E and I to N of Plate 8. Figs. A to E show the course of precipitation when the temperature falls extremely slowly, Figs. I to N when it falls only moderately slowly.[2]

Note that in Figs. I to N the ferrite makes its first appearance at the borders of the austenite grains which have expelled it, as in Fig. I; that the resultant network of ferrite thickens progressively with fall of temperature; and that some of the ferrite which has not completed its journey to the very outside of the austenite grain in which it was born is assembled in the parallel cleavages of the austenite. This last is seen best in the left-hand part of Fig. K and the lower right-hand part in Fig. L.

Throughout this row the time available for the coalescence of the precipitated ferrite into masses visible under these conditions is so short that only the thickest massings, which are in and about the austenite grain boundaries, have thus become visible. Because of this same brevity of opportunity the masses thus thrust into the grain boundaries have retained approximately the position into which they have thus been thrust, and have not yielded to the efforts of surface tension to break these thin networks up into more compact fragments. Hence the marking out of the austenite grain boundaries by the ferrite is very prominent.

In Fig. B a corresponding outlining of the austenite grain boundaries by the ferrite which those grains have expelled to those boundaries is prominent. In Fig. A a corresponding darkening of those boundaries occurs. The reason why these boundaries are dark instead of being white as in the other figures may better be considered later.

But in the remaining figures of C, D, and E of this row the longer time available for the coalescence of the precipitated ferrite, which remains within the grains, into masses visible under these conditions, and for the balling up of these and of the boundary masses under surface tension, results in masking the grain outlines, which under these conditions can at most be guessed.

234. Progress of the Transformation in Rising Temperature Illustrated Micrographically.—Figs. R to Y, Plate 8, show the progress of this transformation when the steel is heated up from j^{VI} to j^{III}, from the mixture of pearlite plus ferrite of which the cold steel consists, to the condition of austenite into which it passes at j^{III}.

Fig. R shows the cold steel, with a well-marked cellular structure, consisting of shells of the lighter ferrite enclosing kernels of the darker pearlite, and with many spines of ferrite shooting into these kernels. Fig. S shows in its left-hand part which has just barely failed to pass j^{V}, the same

[1] Howe and Levy, *Trans. Amer. Inst. Mining Eng.*, 1913, vol. 47, p. 587, and Row 11 of Plate 4, after p. 648.

[2] The Author, *Proc. Amer. Soc. Testing Materials*, 1911, vol. 11, Plate I, after p. 386.

structure as the cold steel of Fig. R. In the right-hand part of Fig. S the steel has just barely risen past j^{V}, and the pearlite of the kernel has changed its color abruptly, showing that before the quenching it had changed from pearlite into austenite; but the ferrite spines and shells remain unlessened in quantity so that no recognizable re-absorption has occurred. In the succeeding Figs., T to V, we see the progressive re-absorption of the ferrite. First the spines weaken and disappear; then the enclosing shells grow thin, become discontinuous, and finally vanish, as the right-hand part of Fig. V.[1]

235. Reason for the Type of Structure which Arises During the Transformation, and Characterizes Slowly Cooled Hypo-eutectoid Steel.—The reason for this structure is readily seen when we consider the conditions under which the ferrite itself comes into existence.

Though the essence of this pro-eutectoid transformation is identical with that of the pro-eutectic solidification, there is the important difference in conditions, that in solidification the pro-eutectic or primary nitrate, cementite, or austenite, forms as a solid within a molten mass, and hence is free to group itself in the form of idiomorphic crystals, whereas in the transformation our present pro-eutectoid ferrite is a solid which is born within the equally solid austenite. True, both ferrite and austenite are relatively soft at this light red heat; yet they are distinctly solid. On this account the ferrite thus formed, thus born, thus generated within the womb and out of the very substance of the mother austenite, has relatively little freedom of motion, little ability to group itself into idiomorphic crystals. In all these cases, whether of solidification or of transformation, that which is solidified or generated is presumably in the form of atoms, or at most of molecular groups, at the instant of its generation. But these atoms or groups thus created within the solid redhot austenite have no freedom of motion, no ability to assemble, comparable with that of the otherwise exactly parallel particles which freeze out of a liquid or molten mass.

Nevertheless they do assemble and coalesce, though incomparably more slowly and into much smaller masses. There is much to suggest that this assembling represents the surface-tension effort of the mother austenite to eject to its outside this new-born ferrite,[2] as it would eject any other foreign

[1] These micrographs represent a single piece of steel, which, after slowly cooling from about 1,400° to the room temperature was cautiously reheated differentially, keeping one end relatively cool, and allowing the other end to rise to a temperature far above j^{III}. It was then quickly withdrawn and quenched in cold water. Thus different points along its length represent heating to different points between j^{V} and j^{III}, and above j^{III}, and the corresponding re-absorption of ferrite. But, though this rapid cooling thus suffices to preserve in a great measure the appearance which the ferrite had at these several temperatures, and though it prevents the return from the state of austenite to that of pearlite which would have occurred had the cooling been slow, yet what it preserves in the kernels is not austenite intact, but martensite and other intermediate stages between austenite and pearlite. See Howe and Levy, *Trans. Amer. Inst. Mining Eng.*, 1913, vol. 47, p. 601, §20.

[2] The Author, *Trans. Amer. Instit. Mining Eng.*, 1913, vol. 47, p. 659. See also *Proc. Amer. Soc. Testing Materials* , 1911, vol. 11, p. 271.

body, as mothers in general and for that matter even the seed vessels of plants, eject their offspring, and somewhat as the eye ejects a grain of sand. Certain it is that the history of the assembling and coalescing of the ferrite particles is approximately that which such a surface-tension expelling effort would cause (§367).

As the eye and not the sand-grain supplies the motive force, so here it is the mother austenite and not the precipitated or new-born ferrite that is the prime mover. This mother metal ejects the ferrite in part to the outside or the boundaries of the various grains of which the austenite itself is composed, and in part assembles it in the octahedral cleavages of those grains. In cases at one extreme, almost the whole of the ferrite is thus ejected completely into the grain boundaries, where it forms the network structure, shown in Fig. D, Plate 1, a white ferrite network enclosing dark meshes of austenite, or of the products of its later transformation. In cases at the other extreme it assembles the ferrite wholly within its own cleavages, giving rise to the Widmanstättian or W structure, shown in Fig. A of Plate 28, a white lattice-work of ferrite within a dark ground mass of austenite, later transformed. But usually the ferrite is distributed in part as an intergranular network, in part within the grains themselves, and much of this latter part lies so irregularly that its connection with the cleavages cannot be detected readily. Note that this assembling of the ferrite is not according to its own crystalline habit, not the self-marshalling of a free body, but a coercion into the cleavages or to the outside of the mother austenite grains, a migration not spontaneous but imposed by the austenite. Evidence supporting this hypothesis is given in §§367 to 370.

An alternative or contributory explanation is based on variations in the concentration or solubility along these surfaces where precipitation occurs.

Note that the expulsion of the pro-eutectoid element to the boundaries of the austenite grains is paralleled by a like expulsion of graphite in making black-heart malleable castings. This is very prominent in Fig. B of Plate 4.

236. Thus the network structure represents an intermediate stage of the general structural development, when the ferrite massed about the austenite grain boundaries has had time to coalesce into visible cell-walls, and before it has had time to obey surface tension so as to spheroidize, and break ranks. Hence the facts (1) that though a very rapid cooling, as by quenching in water, so far restrains coalescence that whatever free ferrite is generated remains invisible; (2) yet a somewhat slower cooling, as in the air, while it gives time for the coalescence in the grain boundaries (shells) to go far enough to mass the ferrite there into visible cell-walls, does not enable the more scattered ferrite within the bodies of the several grains to coalesce into visible masses, with the consequence that the network structure is very striking; and (3) a very slow cooling, as in the furnace, allows the ferrite within the individual grains to coalesce to such an extent as to dis-

tract attention from the cell-walls and to mask them, and allows the cell-walls themselves to break ranks.

What is true of pro-eutectoid ferrite is true of pro-eutectoid cementite. It forms a network in the boundaries of the austenite grains from which it is born, and thus surrounds kernels of the pearlite which result from the final transformation of the austenite in cooling past $Ar1$.

From the fact that both the emergence of the network of the pro-eutectoid element into visibility, and its break up from surface tension and like causes, are the more rapid the greater the quantity of that element,[1] it follows that the farther the carbon content lies from the eutectoid content, the more rapid must the cooling be to catch the development of the structure in this network stage. Steels either slightly hypo-eutectoid or slightly hyper-eutectoid must be cooled slowly in order to give the scanty masses of the pro-eutectoid element time to coalesce into visibility. Steel with less than about 0.40 per cent. of carbon must be cooled rather rapidly to prevent surface tension of the abundant pro-eutectoid ferrite, and the co-operating causes, from concealing the network structure. Hence the network structure is usually associated with steels containing more than about 0.40 per cent. of carbon.

The presence of manganese retards this structural progress to, through, and past the network stage.[2]

As the mass of the ferrite increases and that of the austenite decreases, the surface tension of the ferrite plays a part which gains progressively in importance in comparison with that of the austenite. Thus it comes about that in a very slow cooling, such as is shown in Figs. A to E, Plate 8, the relation of the ferrite masses to the structure of the mother austenite becomes increasingly obscured. It can be traced just above the middle of Fig. E, but elsewhere in this figure it can at best be conjectured. The network structure thus breaks down, partly through balling up because of surface tension, partly through the belated coalescence of finely divided ferrite, partly perhaps through resolution and re-precipitation.[3]

[1] The deforming action of surface tension is the more active the thinner the mass, because the proportion of surface, which is the motive part, to the mass as a whole which the tension has to deform, is inversely as the thickness. Thus the upper part of a falling stream of water holds together because it is thick, but its lower part, moving more rapidly because of the acceleration caused by gravitation, is thinner, and here if anywhere it breaks up into globules or even into spray. This is as true of a thin stream falling from a household faucet as of the mountain cataracts, of which the lower end breaks up into spray, and may even blow away.

But in the present case, the faster ferrite-coalescence in lower than in higher carbon steel results from the more abundant contacts between ferrite masses in the former, because of the greater abundance of those masses, as a thicker milk churns faster than a thinner one, the lower the carbon content the greater being the quantity of ferrite formed in cooling through the transformation range.

[2] The Author, "Life History of Network and Ferrite Grains in Carbon Steel," *Proc. Amer. Soc. Testing Materials*, 1911, vol. 11, pp. 310 and 365.

[3] The Author, *Proc. Amer. Soc. Testing Materials*, 1911, vol. 11, pp. 295 and 355. Howe and Levy, *Internat. Zeits. Metallography*, 1912, vol. 3, p. 4; *Trans. Amer. Instit. Mining Eng.*, 1913, vol. 47, pp. 590 and 591.

Quincke[1] approaches the subject from a wholly different point of view. Without attempting here to explain his theory in full, suffice it to say that he believes that the bounding planes of the austenite grains are analogous to the thin walls in foamy masses of soap bubbles, a honeycomb of austenite slightly different in composition from the remainder of the austenite which it encloses. A simple soap bubble is made nearly spherical by surface tension, and so is the outer or free part of each individual bubble in a frothing mass composed of many bubbles. But the membranous walls between the individual bubbles in such a mass are plane, also because of surface tension, and the skeleton or honeycomb which they collectively form is like that which the bounding planes of the austenite grains form.

His theory is perfectly consistent with the progress of the microstructural changes which I have sketched. However much Quincke may humanly have overworked his theory, the question whether that which determines the grain boundaries is such a membranous surface-tension skeleton, or the position of the nuclei from which growth radiates, whether the bounding planes determine the grain or *vice versa*, or whether each in its degree is a determinant, is not to be settled offhand. The strong curvature,[2] so common among grain boundaries, connects itself more readily with the delimitation of these grains by reciprocal interference than by pre-existing boundaries, such as the straight ones between soap bubbles.

The smoothness and occasional straightness[3] of grain boundaries might suggest that they are self-determining, because their reciprocal interference as they grow toward each other from independent centers, in the way sketched in Fig. 25, might lead to dendritic interlocking, as shown in Fig. 108. But this smoothness may well be due in part to surface tension, and in part to the process of grain growth in cooling from the freezing point down.

237. Martensite. The Hardening of Steel[4] **(§§96 and 228).**—The transformation as we have followed it consists, in cooling down, first of a

[1] Desch gives an excellent summary of Quincke's theories in the *Journ. Inst. Metals*, 1914, vol. 11, No. 1, pp. 57 *et seq.* See also *Internat. Zeits. für Metallographie*, 1913, vol. 4, pp. 23, 79, 303. Earlier papers by Quincke are: *Annalen der Physik*, 1894, [III], vol. 53, p. 593; *idem*, 1902, [IV], vol. 7, pp. 631, 701; *idem*, 1902, [IV], vol. 9, pp. 1, 793, 969; *idem*, 1905, [IV], vol. 18, p. 1; *Proc. Royal Soc.*, 1905, A, vol. 76, p. 431; *idem*, 1907, A, vol. 77, p. 60; *Ber. der deutschen physikalischen Gesellschaft*, 1903, vol. 5, p. 102.

[2] For cases of such strong curvature see Figs. D of Plate 9, A of Plate 14, F of Plate 16, G of Plate 39 of austenite; A of Plate 25, and A of Plate 40 of ferrite; and A of Plate 35 of zinc.

[3] Such straight boundaries are shown, for instance, in cadmium, by Ewing and Rosenhain, *Phil. Trans. Royal Soc.*, 1900, A, vol. 193, Fig. 43, Plate 26, following p. 375.

[4] Hardening by rapid cooling should not be confounded with "case hardening," which is a wholly different operation. Because carbon steel is not given great hardness by rapid cooling unless its carbon content is high, the outer surface of some objects made of low-carbon steel is carburized by heating in contact with carbon, or carbonaceous materials, in order that it may become capable of being hardened thus. This superficial carburized layer is called "the case" and this superficial carburizing is called "case hardening." The hardening power of this case is then made use of by heating the whole object and quenching it in water or oil, which hardens the case intensely while leaving the low-carbon remainder of the object unhardened, and hence with the softness and ductility normal to low-carbon steel. Thus we get the combination of surface hardness to resist abrasion, with malleableness and power of enduring shock on the part of the remainder

progressive change of part of the austenite into pro-eutectoid ferrite in cooling from $A3$ to $A1$, followed by a complete change of all the remaining austenite into pearlite on cooling past $A1$. The change from austenite to ferrite and pearlite thus seems direct, though we noted in §228 the suspected existence of an intermediate beta stage between the austenite or gamma or high temperature and the alpha or pearlite or low-temperature condition.

But if a rapid cooling, for instance by quenching in water as in hardening steel cutting tools, is substituted for a slow one, then more or less of the metal is caught in the intermediate hard martensite stage already noted in §96, a stage which in slow cooling is passed through unnoticed or skipped over according to the point of view.

238. The Essence of Hardening.—The hardening of steel by rapid cooling thus consists essentially in martensitization through the more or less complete suppression of the transformation from austenite into pearlite, and in catching the steel in transit in this hard martensite state. The degree of hardness induced is roughly proportional to the degree in which the metal is caught in the state of martensite, and perhaps to the carbon content of that martensite, probably because the hardness of martensite increases with its carbon content. If the transformation is allowed to go so far that but little of the steel is thus trapped in the martensite stage, most of it having slipped past into the pearlite stage, the hardening is but slight.[1] If on the other hand the transformation is arrested so fully that some austenite persists not yet transformed as far as martensite, then the resultant hardness is less than if this austenite had been allowed to transform as far as martensite; and if the transformation is wholly prevented, the resultant hardness is still further lessened.

239. Acquisition of the Hardening Power.—Thus steel is said to "acquire the hardening power" when its pearlite and ferrite change into austenite on heating up through the transformation range, and to lose that power on again cooling slowly through that range. For if, after it has been heated past the transformation range, it is then allowed to cool slowly, in this slow cooling it passes spontaneously back from the austenite state, through or past the martensite state, and into the state of pearlite plus ferrite, or plus cementite

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of the object. A very rapid carburizing agent is beech or oak charcoal mixed with 40 per cent. of barium carbonate, used at temperatures not above 900°. Wood charcoal alone may be used at temperatures between 950° and 1,000°. Charred leather is used very often, at temperatures between 900° and 950°. (See Brearley, "The Heat Treatment of Tool Steel," Longmans, Green and Co., London, 1911, p. 116; Guillet, "La cementation des aciers au carbone et des aciers speciaux," *Memoires Soc. Ingenieurs Civils*, Feb., 1904, pp. 176 to 207, and *Génie Civil*, 1911, vol. 59, p. 204; Shaw Scott, *Journ. Iron and Steel Inst.*, 1907, No. III, vol. 75, p. 120; Giolitti, "The Cementation of Iron and Steel," translated by Richards and Rouiller, McGraw-Hill Book Co., N. Y., 1915.)

[1] Under these conditions we get neither typical pearlite nor typical martensite. The pearlite is replaced by sorbite, and the martensite by troostite. Sorbite is believed to be only an irresolubly fine emulsion of the constituents of pearlite, and troostite may provisionally be regarded as irresolubly fine martensite.

in the case of hyper-eutectoid steels. The steel once having cooled to below the transformation range, say to 680°, and thus having transformed back past the martensite stage, then no matter how rapidly the remainder of the cooling from 680° down occurs, it has little effect on the hardness of the cold steel. Hence whereas the steel was hardenable when above the transformation range, it ceased to be on cooling to below that range, and remains unhardenable until it rises again past it.

This must needs be the case, for even the most rapid cooling is impotent to return the metal from the soft pearlite state into which it passes in slow cooling, to the martensite state which gives hardened steel its hardness.

Just as rapid cooling hardens steel provided it starts from above the transformation range but not if it starts below that range, so it hardens the steel to an intermediate degree if it starts from within that range, to a degree which represents roughly the proportion of the metal which is caught in transit in the martensite state, and the closely allied state of troostite.[1]

240. Martensitization through Obstruction.—This fact that the transformation from austenite toward pearlite is capable of more or less complete suppression by rapid cooling means first that, though it is rapid, yet under the usual obstructing conditions it is not so rapid that it has time to complete itself in a rapid cooling, and second that it cannot take place in the cold except to a very slight extent.

These obstructing conditions are essentially the presence of one or more of the three retarding elements, carbon, manganese, and nickel. When the quantity of these three elements is very small, as in the case of electrolytic iron and most very low-carbon steel, not even the fastest cooling can prevent permanently the almost complete passage from the high-temperature austenite to the low-temperature ferrite[2] state, with the consequence that no matter how rapidly such metal is cooled, it appears when cold to be free from any important quantity of either austenite or martensite, and to consist of ferrite, with whatever pearlite the small carbon content implies. The steel cannot be hardened greatly. But the presence of a moderate quantity of any or all of these retarding elements so obstructs this transformation that it fails to complete itself during a rapid cooling, with the consequence that thin pieces, cooled rapidly from above the transformation range, consist when cold of martensite, or of a mixture of martensite and austenite, or of austenite alone, according to the rapidity of that cooling and to the quantity of these retarding elements. The steel hardens more or less. Table 34, p. 598, shows how the hardness of rapidly cooled steel increases with the carbon content.

This obstruction increases with the proportion of these obstructing ele-

[1] A hypo-eutectoid steel is more intensely hardened by a rapid cooling from immediately above $A3$ than by one from immediately above $A1$, because of the transfer of the pro-eutectoid ferrite to the state of austenite in rising from $A1$ to $A3$, and for other reasons of detail. See the Author, *Trans. Amer. Instit. of Mining Engineers*, 1913, vol. 45, p. 516.

[2] Because the quantity of pearlite formed in these cases is so small, it would be confusing to call the state into which the austenite transforms in cooling "pearlitic."

ments, and apparently without limit, so that even during a relatively slow cooling the transformation may go only as far as the martensite state, as in the case of the martensitic steels of about 7 per cent. of manganese or about 14 per cent. of nickel, and indeed may not even start, so that the slowly cooled steel consists of austenite, as in the case of 25 per cent. nickel steel and Hadfield's 12 per cent. manganese steel. But carbon alone cannot cause so great an obstruction as this, which is not surprising in view of the very limited quantity of it, apparently about 1.70 per cent., which the austenite is able to dissolve even at the highest temperature practicable. Hence no matter how much carbon is present, in slow cooling the transformation always goes past the martensitic state to the pearlitic state unless a large proportion of manganese or nickel is present.

Guillet, finding that the obstructive power of 1.65 per cent. of carbon if kept in solution is equivalent to that of 29 per cent. of nickel and of 13.5 per cent. of manganese, shows by means of the triangular pyramid of Fig. 30A the limits between the alpha or pearlitic, the martensitic or intermediate, and the gamma or austenitic steels.[1]

The roughly similar obstructing effect of chromium and certain other elements we may here pass by.

241. The Explanation of Hardening.—Let us here survey some prominent theories. Hardening is usually due to martensitization, that is to enabling the transformation to proceed to but not past the martensite stage, either (1) by cooling carbon steels rapidly from the austenite state; or (2) by the presence of such an intermediate proportion of the obstructing elements, carbon, nickel, and manganese, as to cause the transformation to reach only this stage even in slow cooling; or (3) by bringing it to this stage by subcooling cold austenitic steel, *e.g.* in liquid air; or (4) by overstraining austenitic steels by cold deformation, so as to stimulate the overdue change from the austenitic to the martensitic state.

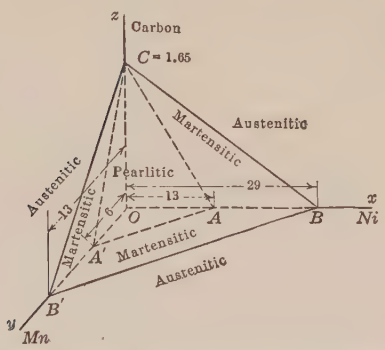

Fig. 30A.—Ratio of the obstructive powers of carbon, nickel, and manganese on the gamma-beta-alpha transformation. (Guillet.)

In addition to these four martensitizing methods of hardening there are

[1] Ordinates measure the content of carbon, abscissæ that of nickel, and distances along the axis $OA'B'$ that of manganese. Alloys the composition of which falls within the pyramid $AA'CO$ are pearlitic or alpha, those between the planes $AA'C$ and $BB'C$ are martensitic, and those in front of the plane $BB'C$ are austenitic, called also gamma or polyhedral. (Discussion of *7th Report Alloys Research Comm., Excerpt Proc. Instit. Mechan. Engin.*, 1905, p. 1031.)

His obstructive-power ratio and his instructive model must be taken as only approximately true. For instance, 1.65 per cent. of carbon cannot retain any iron in the gamma state unless reinforced by rapid cooling, and even then it can retain only a fraction of the iron, whereas 13.5 per cent. and 29 per cent. of manganese and nickel respectively retain the whole of the iron in the gamma state even during slow cooling. Again, his diagram of the iron-manganese alloys differs materially from that of Campbell, Hall, and Howe (U. S. patent 1,113,539, Oct. 13, 1914).

two others, which we may call (5) and (6). (5) is the hardening of austenitic steels by a regulated reheating, which gives mobility enough to allow the transformation to go as far as the hard intermediate state, but not enough to let it reach the final alpha state. This hardening does not always cause martensitization. (6) is the hardening of the malleable metals and alloys in general, gamma and alpha iron included, by plastic deformation. There is neither evidence nor reason for believing that this usually acts through martensitization, except in the case of gamma iron. Let us grasp these six methods firmly.

Though the final states, gamma and alpha, are readily had in their purity, gamma in 25 per cent. nickel steel and in Hadfield's 12 per cent. manganese steel, and alpha in very low-carbon steel, the intermediate state is probably always mixed with a large proportion of either gamma, or alpha, or both, the promptest particles outrunning the rest and reaching the alpha state, the slowest failing to leave the gamma state.

241A. The Reversing and the Cumulative Methods of Hardening.— These six methods fall naturally into two classes, the reversing class which includes (1) to (5), and the cumulative class (6). No matter how far plastic deformation goes, its hardening effect appears to be cumulative without limit.[1]

On the other hand, methods (1) to (5) consist essentially in catching the gamma-alpha transformation in the intermediate hard stage, so that while a full prevention of the transformation gives us the relatively soft gamma state, allowing it to complete itself yields the still softer alpha state. To this I will return.

An additional difference between methods (1) to (5) and method (6) is that the last has a less important effect on the magnetism and electric conductivity than the former class has, and that it affects at least alpha iron much like the other ductile metals and alloys in general, whereas methods (1) to (5) are applicable primarily to the alloys of iron.[2]

Of the five reversing methods, the dynamic ones, (1), (3), and (4) are the hardest to interpret because they leave us in doubt as to how far their phenomena are due to the stress, strain, and amorphism to be expected from the dynamic conditions, and how far to the common cause which we naturally infer to underlie all five methods, from the fact that the hardness given by the static methods is nearly identical with that given by the dynamic ones. Therefore our attention should be fixed primarily on the static methods.

[1] It is true that because our actual methods determine the hardness not of the ultimate particles but of the mass, the cracks which excessive deformation causes, by lessening the coherence of the mass, lessen the apparent hardness. So, too, deformation may induce internal stresses so severe as to lessen the resisting power of the mass to external stresses, the power which is actually measured by our hardness methods, and may thus lessen the apparent hardness of the mass. Yet there is no reason to suspect that the hardening effect of deformation on the ultimate particles is limited and reversible.

[2] The hardening which certain copper alloys undergo when cooled rapidly from above their transformation range will be considered in §252.

From them we can proceed to consider the effects which the dynamic conditions superpose on the hardening effect of the common cause.

241B. Examination of the Reversing Methods of Hardening.—Under static conditions, as when determinations are made at stationary temperature on pure iron, the passage from the gamma to the intermediate state at $Ar3$ is accompanied by a great expansion (Fig. 31, I). At a lower temperature, $Ar2$, there are at least as striking evidences of transformation in the rapid recovery of the magnetic properties, and a sharp break in the temperature-electric conductivity curve (Fig. 31, II), with but slight change in the dilatation curve (Fig. 31, I), and relatively slight heat evolution. With further cooling there is at first a continuous gain of magnetism and a continuous change in electric conductivity.

The volume changes are shown by the dilatation curve of pure iron *EFDLM*, Fig. 31, I, by Benedicks. From *E* to *D*, that is from a point far below $A2$ to $A3$, the dilatation follows a nearly smooth curve, in which the greatest discontinuity which can represent $A2$ is less than $^3/_{1000}$ of the break at $A3$ according to Benedicks. The extremely sensitive thermal determinations of Burgess and Crowe showed only a moderate break at $A2$, 768°.[1]

To fix our ideas let us call these steps 1 and 2 respectively, step 1 being from the gamma to the intermediate state at $Ar3$, accompanied by great expansion but not by recovery of magnetism, and step 2 being from the intermediate to the alpha state at $Ar2$, accompanied by great recovery of magnetism but by little change in volume. For our present general survey we need not ask whether $Ar2$ represents the whole of step 2 or only an important part of it, the rest being spread out, possibly over a wide range of temperature.

In the presence of much nickel and manganese it appears to be spread out very widely.

These two steps are indicated roughly in Fig. 31, VII. We shall see that step 1 is accompanied by great hardening[2] and step 2 by great softening. Thus step 1 is from the relatively soft, dense, non-magnetic gamma state to an intensely hard, bulky, but non-magnetic one. The second step is to the final bulky, very soft, and very magnetic alpha state. Note carefully that the changes in magnetism and volume do not correspond to those in hardness.

[1] "The Critical Ranges $A2$ and $A3$ of Pure Iron." *Trans. Amer. Inst. Mining Eng.*, 1913, vol. 47, p. 665.

H. C. H. Carpenter, "The Critical Ranges of Pure Iron with Special Reference to the $A2$ Inversion" (*Journ. Iron and Steel Inst.*, 1913, No. I, vol. 87, p. 315), could find no discontinuity at all in this range in the heating and cooling curves. His measurements, though very sensitive, were less sensitive than those of Burgess and Crowe.

[2] Greatly as the intermediate state exceeds the gamma and alpha states in hardness at the room temperature, at the high range of temperature, $A2$, at which it is in equilibrium in pure iron it is but little harder than alpha iron, and apparently softer than gamma, if we allow for the natural softening effect of the somewhat higher temperature, $A3$, at which gamma is in equilibrium. (Rosenhain and Humfrey, *Journ. Iron and Steel Inst.*, 1913, No. I, vol. 87, p. 247.) Hence when its great hardness is referred to, hardness in the cold is to be understood. It hardly surprises us that the relative hardness of these three states varies thus greatly with the temperature.

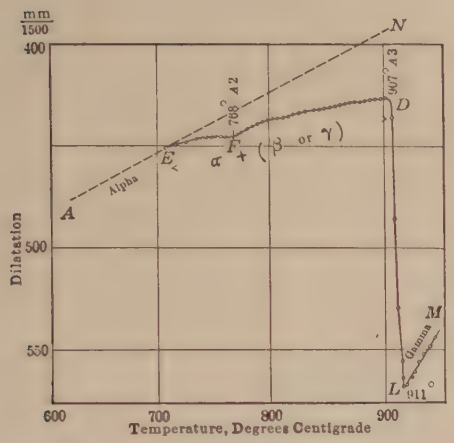

FIG. 31, I.—The dilatation of pure iron in heating, after Benedicks.

NOTE.—The line AEN is sketched in.

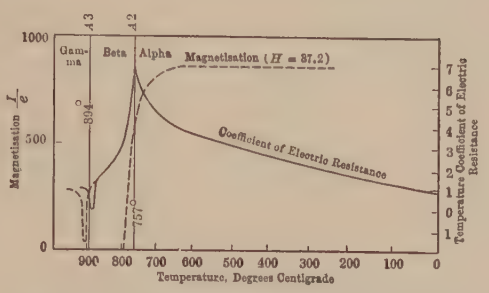

FIG. 31, II.—Variations in the magnetic permeability and temperature coefficient of electric resistance of pure iron, with the temperature. (Burgess and Kellberg, and Honda.)

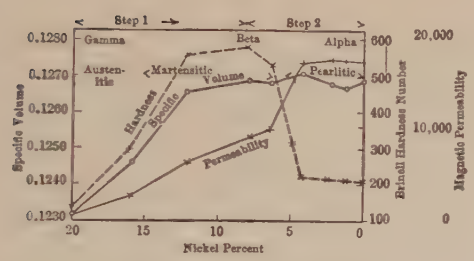

FIG. 31, III.—Variations of the microstructure and physical properties with decreasing nickel content. (Carpenter, Hadfield, and Longmuir.)

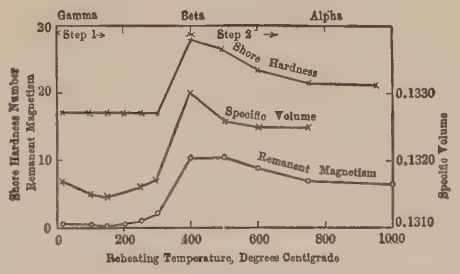

FIG. 31, IV.—Increase of hardness, volume, and remanent magnetism on reheating 2 per cent. manganese austenite. (Maurer.)

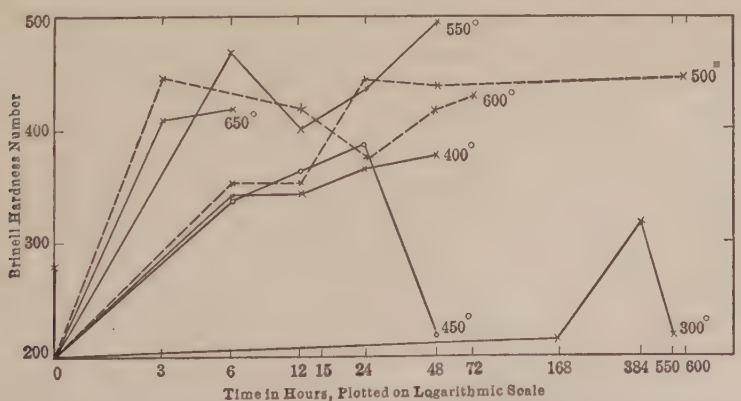

FIG. 31, V.—Influence of the time and temperature of holding the manganese steel of Hopkinson and Hadfield on its hardness.

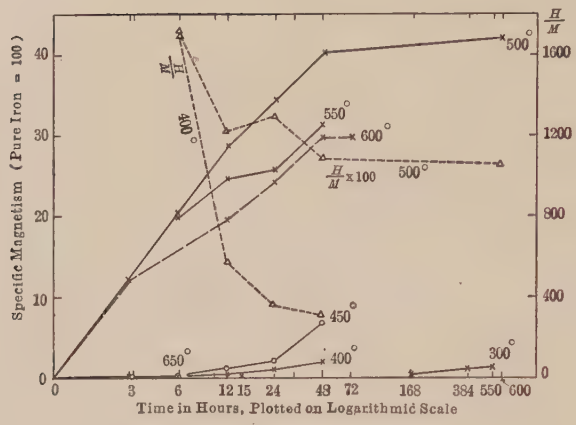

Fig. 31, VI.

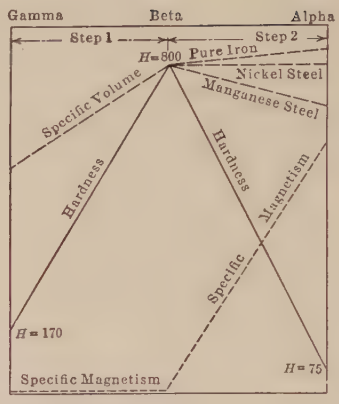

Fig. 31, VII.

Fig. 31, VI.—The specific magnetism of the manganese steel of Hopkinson and Hadfield increases with the length of holding at given temperature. For given length of holding period, it increases with the temperature up to 500°, and thenceforth decreases.

Dotted lines represent the ratio of the hardness, H, shown in Fig. V, to the magnetism, M, here shown after reheating for different periods to 400° and 600°.

Fig. 31, VII.—The two typical steps in the gamma-beta-alpha transformation.

Note to Fig. 31.—The data underlying Figs. 31, I to VI severally are from the following sources:

I from Benedicks, *Journ. Iron and Steel Instit.*, 1914, No. I, vol. 89, Fig. 7, after p. 440.

II from Honda, *idem*, 1915, No. I, vol. 91, p. 212, Fig. 5a, and *Science Reports, Tôhoku Imper. Univ.*, 1913, vol. 2, No. 2, Fig. 11a, after p. 94; and Burgess and Kellberg, *Journ. Washington Acad. Science*, 1914, vol. 4, Fig. 3, p. 439.

III from Carpenter, Hadfield, and Longmuir, *7th Report Alloys Research Comm., Excerpt Proc. Instit. Mechan. Engin.*, 1905, pp. 878–881 and 904.

IV from Maurer, *Rev. Metallurgie, Mem.*, 1908, vol. 5, pp. 721, 726, 732; and *Metallurgie*, 1909, vol. 6, pp. 38, 41, and 43.

V and VI from Hopkinson and Hadfield, *Trans. Amer. Instit. Mining Engin.*, 1914, vol. 50, p. 482; *Journ. Iron and Steel Instit.*, 1914, No. I, vol. 89, p. 117.

Thus in step 1 there is a great gain in hardness and volume but not in magnetism, and in step 2 the great loss of hardness is accompanied by a great gain in magnetism but by little change in volume.[1]

Of the three states, one only, gamma, is dense, one only is hard, the intermediate one, and one only, alpha, is magnetic.

Method (2). Under other static conditions, as when we pass in the cold from the gamma through the intermediate to the alpha state in examining a series of slowly cooled alloys of progressively decreasing nickel, content Fig. 31, III, a passage equivalent to method (2), step 1 from Ni about 20 to Ni

[1] Maurer's 2 per cent. manganese austenite, Fig. 31, IV, in passing from the hard intermediate to the soft alpha state lost about one-third of the gain of volume which it underwent in changing from the gamma to the intermediate state. On the other hand, in the series of Carpenter, Hadfield, and Longmuir, Fig. 31, III, though in passing from the soft gamma state with 19.91 per cent. of nickel to the martensitic state with from about 13 to about 6 per cent., there was a great expansion (loss of density), in passing beyond to the end of the series, with no nickel, the volume changes are small enough to be explained by the changes in the quantity of cementite and by observational errors, assuming that with percentages of nickel less than 5, a small quantity of allotropic iron may persist, and thus change the percentage of alpha and hence of free cementite.

In pure iron, Fig. 31, I, the fact that the curve is flatter from D to E than from M to L and probably than from E to A suggests that here some expansion accompanies the beta-alpha change.

about 12 per cent. is accompanied by a like expansion, and by a change from the gamma or polyhedral microstructure to that of martensite, but with relatively slight recovery of magnetism; and step 2, from Ni about 8 to Ni 4, by a great recovery of magnetism with but slight volume changes. Moreover, and this is the important thing, step 1 is accompanied by an increase of hardness from 202 in the gamma to 578 in the intermediate martensitic state; and step 2 by a fall of hardness to 131,[1] or better to 75.

Method (5). Under still other static conditions, as when gamma iron is carried through divers stages of these two steps by heating it to a series of temperatures high enough to allow the transformation to proceed to different limited degrees, Fig. 31, IV, step 1 is accompanied by great hardening and expansion and by a certain recovery of magnetism, and step 2 by further recovery of magnetism, great loss of hardness, and a great change in electric conductivity, but by relatively slight volume changes.[2] Thus the changes are roughly parallel to those in pure iron examined at different temperatures, Figs. 31, I and II, plus the gain of hardness by step 1 and its loss by step 2, though how much of the magnetism is recovered in each step is not clear.

Method (3). Under relatively simple dynamic conditions, step 1 (taken by cooling gamma iron in liquid air), is accompanied, as in the static cases,

[1] Carpenter, Hadfield, and Longmuir, *loc. cit.*, pp. 863, 880, and 904.

This number 131 takes no account of the hardening effect of the introduction of cementite which accompanied step 2 in this specific case. What we really want to compare with the hardness of the gamma and of the intermediate state, which are free from cementite, is that of the alpha iron itself, which is probably not far from 75. Moreover, the hardness of gamma in its purity is probably less than the 202 here reported, or than the 170 which I found, because the deformation incidental to the hardness determination itself shifts some of the iron from the initial soft gamma state to the intermediate state, by method (4), so that the hardness which we determine is not that of the initial gamma iron, but of the mixture of gamma plus intermediate developed in the course of the hardness test itself. In §644 I estimate it as not greater than 125.

In these experiments alloys containing the percentages of nickel indicated, and about 1 per cent. of manganese and 0.5 per cent. of carbon, were normalized by heating to somewhat above their transformation range, 800°, and cooled in about 16 hours to room temperature. With 19.91 per cent. of nickel the structure was gamma or polyhedral; with between 15.98 and 6.42 per cent. it was martensitic; with less it was pearlitic or alpha.

[2] Maurer, *Rev. Metallurgie, Mem.*, 1908, vol. 5, p. 711, and *Metallurgie*, 1909, vol. 6, p. 33; Hadfield and Hopkinson, *Trans. Amer. Inst. Mining Eng.*, 1914, vol. 50, p. 476, and *Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, p. 106; Sauveur, *Trans. Amer. Inst. Mining Eng.*, 1914, vol. 50, p. 501; the Author, *Trans. Faraday Soc.*, 1915, vol. 10, p. 267.

Maurer, treating steels of 0.93 C with 1.67 Mn and of 1.94 C with 2.20 Mn, converted them into "homogeneous austenite" by quenching them from 1,050°. On reheating this so as to allow the arrested transformation to proceed, the changes shown in Fig. 31, IV occurred. The "remanent" magnetism is given. The sharp increase of this at 400° along with the change in hardness, density, and conductivity, shows that some alpha iron formed, but that the quantity of non-magnetic iron was still great. The decrease of the remanence from 400° to 800° implies the progressive transfer of the allotropic non-magnetic to the magnetic or alpha state, the magnetic retentivity implying the coexistence of magnetizable alpha iron with inert and hence unmagnetizable and hence allotropic iron. Thus whereas great permeability implies a very large proportion of alpha iron, great remanence implies an intermediate quantity, mixed with an intermediate quantity of allotropic iron.

The results of Hadfield and Hopkinson, shown in Figs. 31, V and VI, support Maurer's, as will be explained in §244.

by an increase in hardness and in volume, and by a change from the polyhedral gamma structure to that of martensite.[1]

Method (4). The same results follow taking step 1 under the relatively simple dynamic conditions of deforming gamma iron in the cold.[2]

Though methods (4) and (5) can be managed so that the transformation is confined almost wholly to step 1, so that great hardness results with but little magnetism, hardness 540 with specific magnetism only 0.8 by method (4) and hardness 418 with specific magnetism only 0.2 by method (5),[3] methods (2), (3), and (4) have thus far resulted in allowing more or less of step 2 to accompany step 1, so that more or less alpha iron and consequent magnetism accompany the intermediate state to which step 1 in its purity would lead.

[1] Hopkinson (*Proc. Royal Soc.*, 1890, vol. 47, p. 138, *Journ. Iron and Steel Inst.*, 1896, No. I, vol. 49, p. 208) found that 25 per cent. nickel steel which was non-magnetic at the room temperature, when immersed in carbonic acid snow became magnetic, lost in electrical resistance, and gained in hardness and volume, the specific gravity falling from 8.15 to 7.98. Le Chatelier (*Comptes rendus*, 1890, vol. 110, p. 283), brought about the same transformation. Osmond (*The Metallographist*, 1899, vol. 2, p. 261, and Osmond and Stead, "Microscopic Analysis of Metals, 1913, p. 191), prepared a mixture of austenite and martensite by quenching blister steel containing about 1.50 per cent. of carbon from about 1,050°. The austenite as usual was softer than the martensite. On subcooling this in liquid air its magnetic permeability and remanent magnetism increased, and it gained in volume, the specific gravity falling from 7.798 to 7.692. On the previously polished surface of the bar this subcooling caused a martensitic pattern to appear. At the same time the hardness of the austenitic part of the specimen increased, though it did not reach that of the martensitic part. Benedicks (*Journ. Iron and Steel Inst.*, 1908, No. II, vol. 77, p. 250), quenched steel of 1.33 per cent. and 1.99 per cent. of carbon from 1,050°, when it contained a mixture of austenite and martensite. On cooling it in liquid air and then removing it into the air it emitted clicks. Here the change from austenite to martensite is to be referred to the release of pressure due to the sudden expansion of the outer layers on passing from the liquid air to the room temperature, permitting the expanding gamma-beta change to occur. He further found (*idem*, p. 245) that the austenite is absent from the very surface of quenched specimens, which is in tension and therefore transforms as far as martensite, though it may be present in the very interiors where the pressure prevents the expansion necessary to the formation of martensite. In the same way he found that the release of pressure caused by grinding the surface from a quenched specimen caused martensite to form and to show its characteristic markings on the surface thus polished. Further, increasing the pressure during quenching increases the retention of austensite. Hence it is probable that, in all these cases of martensitization and hardening by subcooling or by deformation, the true cause is the sudden release of pressure during the subsequent rise of temperature or by the deformation.

[2] See Guillet, *The Metallographist*, Feb., 1906, vol. 11, p. 92. Cold working, that is plastic deformation or overstrain, martensitizes those austenitic steels the composition of which approaches the limit beyond which they are martensitic when cooled slowly. This martensitization is accompanied by the usual increase of volume. See §642.

Maurer (*Rev. de Metallurgie*, 1908, vol. 5, p. 744), turned his homogeneous austenite of carbon 1.94, manganese 2.24 per cent. into martensite by either scratching it or striking it with a hammer.

Carpenter, Hadfield, and Longmuir (*7th Report Alloys Research Comm.*, 1905, pp. 949 and 953, Plate 57, Figs. 55 to 57,) found that deforming their 19.91 per cent. austenitic nickel steel turned it into martensite and made it magnetic. In machining this steel it was soft "at the very beginning" (p. 949) "and up to a certain point easily machinable (p. 863), but quickly became hard and martensitic. Tensile rupture increased the hardness of Hadfield's manganese steel from presumably about 220 to 540, with a specific magnetism of 0.8 in the experiments of Hadfield and Hopkinson (*Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, p. 112), and to 340 in my own experiments, the increase of hardness being five times in the former case and twice in the latter case that which I found that low-carbon steel undergoes in tensile rupture. (See §642.)

[3] Hadfield and Hopkinson, *Trans. Amer. Inst. Mining Eng.*, 1914, vol. 50, pp. 482–3, and *Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, p. 112.

Method (1), the most dynamic of all, the hardening of carbon steel by quenching, allows a still larger fraction of step 2 to accompany step 1, so that much alpha iron and hence much magnetism accompany the hardened or intermediate stage to which step 1, as taken in this process, leads. Nevertheless, here too the characteristics of step 1 are prominent, a great increase of hardness and in volume, with an imperfect recovery of magnetism.

242. Hardening is Allotropic.—What is the nature of the intermediate hardened state which these five methods cause in common? The agreement is almost universal that it is allotropic, in the sense that the iron of the hardened steel is not the alpha iron familiar to us in our common annealed steel. Opinion is divided as to the nature of this allotropy.

We know three allotropic states of iron, (1) the common low-temperature, alpha, or pearlitic state; (2) the high-temperature, gamma, or austenitic state; and (3) the amorphous state[1] brought about (A) by the shattering of the individual crystalline particles at a temperature too low to permit their reassembling in crystalline form, as in cold rolling, wire-drawing, and other forms of overstrain; or (B) locally at grain junctions where the space between abutting differently oriented grains is narrower than the smallest possible crystal unit, so that crystallization of this residual parting is impracticable. The intermediate state reached by step 1 also may be allotropic, beta iron, normal to a temperature range between the gamma and the alpha regions, and familiarly represented by the triangle *GOM* in Fig. 23. In steels at the right of *O*, both hypo- and hyper-eutectoid ones, this triangle is continued by the lines *OS*, and *SK*, that is to say in these steels the beta state is passed through on crossing *OS* or *Ar*3–2 in the case of hypo-eutectoid steels, and *SK* or *Ar*3–2–1 in the case of hyper-eutectoid steels. Whether this condition is passed through or skipped over in slow cooling, it is, on the beta iron hypothesis, passed into in the rapid cooling of carbon steel, and also in the four other methods before us (2) to (5).

In iron alloyed with from 6.42 to 15.98 per cent. of nickel there is indeed a wide range of temperature in which martensite, if not strictly in equilibrium, at least has such great stability that the steel remains martensitic whether cooled slowly or rapidly, though the poorer of these alloys can indeed be shifted over at least in large part into the pearlitic state by a 50-hour stay at about 140°, which is near the lower limit of the transformation range for their nickel content. Yet none of them are readily caught in the gamma state. The width of this martensitic range in the presence of much manganese or nickel contrasts strongly with the apparently mathematical line, *OSK* of Fig. 23, which seems to correspond to the martensitic state in carbon steels at the right of *O*.

243. Three theories of the nature of martensite and thus of the hardening of steel may be considered here, the beta, the gamma, and the amor-

[1] I trust that those who do not class the change caused by plastic deformation as allotropic will, on the ground of convenience, pardon my using allotropy in this broad sense. If with Beilby we hold that this change consists in making some of the metal amorphous we reasonably call it allotropic.

phous, which refer the martensitic state and structure and the accompanying hardness to the presence of beta iron, of a solution of gamma in alpha iron, and of amorphous iron respectively.

244. On the beta theory (§228), in the rapid cooling of carbon steel the transformation is so far obstructed that, though part of the iron goes as far as the alpha state, some of it is caught in transit in the beta state. Again the presence of much manganese or nickel so far retards the transformation, partly directly and partly through lowering the normal temperature at which the transformation is due, that in like manner it remains very incomplete, much iron being caught in transit in the beta state.

So with the four other reversing methods of hardening (2) to (5). They all consist essentially in allowing or inducing the transformation to go from the austenitic or gamma state as far as the martensitic or beta state.

The discontinuities at $A2$ in the curves of magnetism[1] and electric conductivity Fig. 31, II, indicate that a profound change in the nature of the metal occurs at this temperature. The spreading out or the concentration of this change, the exactness or inexactness of its reversibility, and the fact that it causes only slight volume changes, neither affect in the slightest the profundity of the change nor offer good reason for attempting to alter arbitrarily the meaning of allotropy with the purpose of excluding this change from the allotropic ones, in the present shallow stage of our knowledge of allotropy. Should it become expedient to classify the allotropic changes in any desired manner, whether into the molecular and the atomic or otherwise, the proper procedure is, not to change the established meaning of the generic name "allotropy," but to propose specific names for its subdivisions.

The evidence seems irresistible that, in the cooling of pure iron, a profound change from a non-magnetic to a magnetic state, or better from a far less to a far more magnetic state, occurs at $Ar2$, with slight and outspread volume changes; that this change is radically distinct from the change accompanied by the great expansion without recovery of magnetism which occurs at $Ar3$; and thus that the transformation in the iron itself consists of two distinct steps, step 1 from dense non-magnetic gamma to bulky non-magnetic intermediate at $A3$, and step 2 from bulky non-magnetic intermediate to bulky magnetic at $A2$.

The presence of an abundance of the obstructing elements, especially of nickel or of manganese, enables us to isolate the intermediate state, not indeed in its purity, but in such concentrated form that its properties shine forth. We have seen four methods (2) to (5), which thus enable us to isolate it far more completely than by quenching carbon steel. All four give as concordant results as could be expected in view of the great differences in

[1] See Benedicks, *Journ. Iron and Steel Inst.*, 1912, No. II, vol. p. 86, 261.

P. Curie, *Annal. chim. phys.*, 1895, ser. 4, vol. 5, p. 289; and "Oeuvres," Gauthiers-Villars, Paris, 1908, p. 232.

Weiss and Foëx, *Archives des sciences*, Geneva, 1911, ser. 4, vol. 31, p. 89.

K. Honda, Science Reports, Tôhoku Imp. University, 1913, vol. 2, p. 69; *Journ. Iron and Steel Inst.*, 1915, No. I, vol. 91, p. 242.

the completeness of the isolation of the intermediate state. They all show the intermediate to be incomparably harder than either gamma or alpha iron. From the fact that the mixture of intermediate with alpha in the 7.05 per cent. nickel steel of Fig. 31, III, reached in the absence of quenching stresses by the purely static method (5), though it has enough alpha to give very considerable magnetism, yet has a hardness of 578, seems to justify the inference that, in the complete absence of both gamma and alpha, the hardness of the intermediate state would be at least 750 if not 800, or 4.4 times that of the gamma iron of Hadfield's manganese steel and 10 times that of alpha iron.

To such an intermediate allotropic state, far more bulky than one of its neighbors, incomparably less magnetic than the other, and incomparably harder than either, the name beta is wholly appropriate, and by that name I shall call it unless in discussing the amorphous and the solution hypotheses which have been offered as substitutes for the beta hypothesis.

The fact that the specific magnetism of carbon steel hardened by rapid cooling, though it is small in weak magnetic fields, is great in very strong fields,[1] has been interpreted as opposed to both the beta and the gamma theory, on the ground that, if the hardening were due to the presence of allotropic iron, either beta or gamma iron, then because gamma iron, and beta iron if it exists, are non-magnetic at high temperatures, if they were present abundantly in the hardened steel they would preclude great specific magnetism. But this is a fallacy. The difference in magnetic behavior between the iron of hardened and that of unhardened carbon steel is quite great enough to justify belief that the hardened steel contains much allotropic iron. The difference as regards specific magnetism in weak fields is very great, and this difference suffices. We might and generally did expect that a like difference would be found in strong fields, but this latter difference is in no way a necessary result of allotropy, nor is its absence a true suggestion against the existence of allotropy. It is always to be remembered that different allotropic forms may be identical in every respect but one, and that a difference in one property, if surely proved, is in itself proof of allotropy. Of course, when the number of properties in which such differences occur is extremely small, we must demand all the more conclusive and abundant evidence that they really exist. For that matter the belief that beta and gamma are non-magnetic when hot does not imply that they are when cold.

Hence, the magnetism of hardened steel differing markedly from that of the unhardened steel in weak fields, even their complete identity in strong fields would not really disprove the allotropy of the iron of the hardened steel. Nevertheless the proof, by Hadfield and Hopkinson, that

[1] Hadfield and Hopkinson (*Proc. Inst. Electrical Engineers*, 1911, vol. 46, part 206, p. 235) find that the specific magnetism of steel of 0.85 per cent. of carbon when hardened is 89 per cent. of that of unhardened steel in an intense field, though in moderate fields this percentage is much smaller. See note B, p. 624.

in strong fields the intensity of magnetization of quenched high-carbon steel, with a Brinell hardness of about 640, lacks only about 11 per cent. of equalling that of ferrite with hardness of about 75, if it implies that only 11 per cent. of the mass of the hardened steel is non-magnetic, indicates that in addition to the non-magnetic beta iron as we know it above $A2$ and in the hard and non-magnetic manganese and nickel steels of Figs. 31, III to VI, there is a fourth allotropic form, hard and yet magnetizable in strong fields, which we may call beta II. On this hypothesis, step 2 of the transformation is really divisible into two parts as sketched in Fig. 31, VIII, so that the hard state of Figs. 31, III to VI represents beta I mixed with gamma, and that of quenched carbon steel represents beta II mixed with alpha. This hypothesis tallies with the suggestion of Fig. 31, II, that the extinction of magnetism is at a temperature above that of $A2$ as indicated by the electric resistance; with the irregularities between D and E in Fig. 31, I; and with the frequent doubling of the thermal point $A2$.

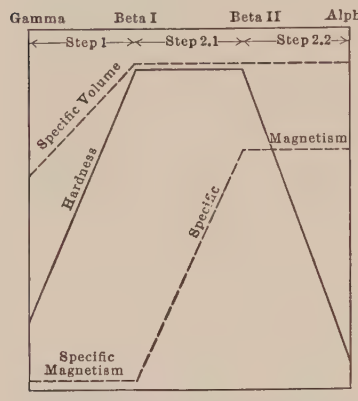

FIG. 31, VIII.—Steps in the gamma—beta I—beta II—alpha transformation. See note C, p. 624.

It tallies with the conception that this intermediate state is allotropic, that by whichever of the five very different methods, (1) to (5), some purely static, one even violently dynamic, the intermediate stage is reached, it has its striking property of hardness far in excess of that of either gamma or alpha iron, the excess of its hardness and its degree of magnetism being capable of reconciliation with the quantity of gamma and alpha which may be expected to coexist with it. Thus method (2), in bringing about a progressive transfer from the gamma through the beta to the alpha states, while it gives a great maximum of hardness along with the martensitic state, has a progressive increase of magnetism as the proportion of alpha presumably increases progressively. The subcooling of steel already martensitic in large part and only slightly austenitic, method (3), in enabling the arrested transformation to proceed a little further causes some swelling, and a moderate increase in hardness and in magnetism.

245. The gamma theory differs from the beta only in holding that the intermediate martensitic state thus reached is not a distinct beta state, but merely a solution of gamma in alpha, essentially because there is no important break in the curve of dilatation in the supposed beta region, say ED of Fig. 31, I, such as might be expected if there were a change from a low-temperature to a high-temperature allotropic state here. Hence the gamma theory denies the existence of beta iron, and, as the essence of the martensite state, substitutes for the supposed beta iron a supposed solution of gamma iron in alpha iron in varying proportions.

To this the beta hypothesis replies, first that the breaks in the magnetic

and conductivity curves at $A2$ are far too sharp to be referred to simple changes in the ratio in which gamma is dissolved in alpha; second that, in view of the sharp breaks in these curves, the flatness of the thermal and volume curves is to be interpreted as meaning that the beta-alpha change is accompanied by only slight thermal and volume changes, and that indeed the discontinuities in these curves at about $A2$ may be partly or even wholly due to changes in the nature of the transformation. That is to say, just below $Ar3$ the metal may consist of a mixture of gamma and beta I, and and as the temperature falls slowly, step 1 may continue and steps 2.1 and 2.2 may be taken in ratios which vary continuously and so as to cause the irregularities which occur between D and E of Fig. 31, I.

The beta hypothesis replies further and reasonably that the hardness which always accompanies the martensitic state, and indeed the martensitic structure itself, are explained more easily as due to the existence of a hard allotropic beta iron, than as due to the solution of one soft allotropic form, gamma, in another soft one, alpha. Freely admitting that a solution may be somewhat harder and more brittle than either of its constituents, we may hold reasonably that the excess of the hardness of martensitic steel over that of both alpha pearlitic and gamma austenitic steel is too great to be explained readily by the mere solution of gamma in alpha, and that this is even more true of the brittleness. Martensite is habitually perfectly brittle, and such perfect brittleness is an antecedently improbable result of dissolving one very ductile allotropic form, gamma iron, in another also very ductile one, alpha iron.[1]

246. Additional Evidence for the Beta Hypothesis.—Cogent evidence for the beta as against the gamma hypothesis is the great variation in the ratio between the hardening and the magnetism induced by arresting the trans-

[1] Manganiferous austenite of 0.85 per cent. of carbon will sometimes elongate 50 per cent. in 8 in. (the Author, "The Metallurgy of Steel," 1890, p. 361), and moreover is so ductile that cracks once started in it can be propagated only with great difficulty (the Author, *Journ. Franklin Inst.*, Feb. 20, 1892). Pure ferrite probably has an elongation of at least 40 per cent. in 8 in., for Cushman reports an elongation of 38.8 per cent. in 8 in. in A.R.M. steel. (*Proc. Amer. Soc. Testing Materials*, 1911, vol. 11, p. 406.) It is doubtful whether hardened steel of 0.85 per cent. of carbon, if in thin enough pieces to harden fully, would undergo any elongation at all before breaking. Still more striking is the combination of an elongation of 75 per cent. in 2 in. with a tensile strength of 112,000 lb. per square inch reported by Hadfield in austenite containing C 0.60, Ni 14.55, and Mn 5.04 (Engineering conference, *Proc. Inst. Civ. Engineers*, Supplement to vol. 154, p. 118, and *7th Report Alloys Research Comm., Excerpt Proc. Inst. Mechan. Engin.*, 1905, p. 873). (§§642 to 645.)

The intrinsic improbability of the gamma hypothesis is seen by applying it to two familiar cases. Hadfield's quenched manganese steel is a solution of some 19 to 25 per cent. of cementite in gamma iron, undercooled at the room temperature by some 625°. On the gamma hypothesis quenched carbon steel of say 0.40 per cent. of carbon is a like solution in gamma iron undercooled by about 700°, which in addition to 6 per cent. of cementite contains also much alpha iron in solution. On this hypothesis it differs in constitution from Hadfield's steel in that alpha iron is substituted for some (19 to 25) − 6 = (13 to 19) per cent. of cementite. Hadfield's is very ductile, quenched carbon steel is glass brittle. On this hypothesis, then, to replace in the solute a large quantity of glass-brittle cementite with extremely ductile alpha iron changes the resultant solution from great ductility to glass brittleness, the solution with the ductile alpha iron being glass brittle, and that with much glass-brittle cementite being extremely ductile. (The Author, *Trans. Faraday Soc.*, 1915, vol. 10, p. 268.)

formation under different conditions. If, following the gamma hypothesis, the progressive changes in hardness and the progressive recovery of magnetism simply represent the progressive substitution of alpha for gamma, the transformation being direct from one phase to a second, then a given degree of hardness in given specimens should represent at most two stages in that transformation. That is to say, as, with the supposed progressive substitution of alpha for gamma iron in this solution the hardness rises to a maximum and then decreases, any given hardness should exist first in the passage up to that maximum and second in the passage down from it, and should in each case correspond to one approximately fixed ratio of gamma to alpha. To each of those ratios an approximately fixed degree of magnetism should correspond, so that a given degree of hardness should, under varying conditions, be accompanied by at most two relatively fixed degrees of magnetism.

But if, following the beta hypothesis, the transformation is from gamma through beta to alpha, being a plural step transformation involving three or four phases—one dense, one or two hard, and one or two magnetic—then by varying the ratio in which these steps are taken, we might be able to set up any conceivable combination of hardness and magnetism. The taking of step 1 by a given number of particles, resulting in a given hardness, might be accompanied by much or little magnetism according to whether many or few particles simultaneously took step 2 or 2.1 and so became magnetic.

This is exactly what occurs. In the data of Hadfield and Hopkinson I find that in reheating gamma iron and thus enabling the transformation to proceed by method (5), a given hardness, 418 to 420, may be accompanied by a specific magnetism of 0.2, or 0.19, or 30. With slightly wider hardness limits, 418 to 438, the specific magnetism may be 0.2, or 19, or 25.8, or 29.8, or 30, in each case a 150-fold increase in magnetism, with intermediate stages. Moreover the time and temperature relation is just what the beta hypothesis calls for, the magnetism increasing with the time while the hardness remains almost constant, as shown in Figs. 31, V and VI.[1] Clearly in

[1] Hadfield and Hopkinson brought Hadfield's 12 per cent. manganese steel to the non-magnetic gamma state by quenching it from about 1,100°. They then reheated it to many temperatures and for many periods. The results plotted in Figs. 31, V and VI show that, at each temperature, whereas the gain of hardness very soon reached its maximum, showing that step 1 was taken quickly, the gain of magnetism occurred extremely slowly, showing that step 2 not only is separate from step 1, but even lags far behind it. Sauveur, discussing these data and adding to them, showed that they accord with the beta theory. I added further confirmatory data (*Trans. Faraday Soc.*, 1915, vol. 10, p. 267).

The hardening caused by step 1 is here exaggerated somewhat by the precipitation of the proeutectoid cementite which occurs on heating this form of austenite. But there is good evidence that this can account for only a part of the great gain in hardness caused by this method (5), in one case from about 175 to 495, or by 320 numbers. A cooling slow enough to precipitate most of the cementite in this steel increased the hardness in my experiments by only 20 numbers, from 170 to 190 (slow cooling in the sand from the molten state), and by only 35 numbers, from 177 to 212, in Sauveur's experiments (substituting a furnace for a water cooling), one-sixteenth and one-ninth respectively of the gain by method (5).

Note particularly that, while the gain of magnetism is progressive, continuing at 500° even after 48 hours and at 300° even after 384 hours, the hardness reaches an approximate maximum at the

the early part of the heating many particles take step 1, but few take step 2. As time goes on more and more take step 2, and the magnetism increases progressively, the proportion of beta, and hence the hardness, being held approximately constant by fresh transfers of gamma to beta. Again, Hadfield[1] finds that method (4) may give a hardness of 540 with a specific magnetism of only 0.8, indicating that step 1 to the hard beta state has been taken by a large proportion of the particles, but step 2 to the magnetic alpha by very few. Here some allowance must be made for the amorphism which this method causes incidentally.

Striking support for the beta hypothesis comes here in the much greater ratio of magnetism to hardness in the hardening of carbon steel than in this Hadfield and Hopkinson hardening by method (5). The specific magnetism of hardened quenched carbon steel of hardness 420 is about 95, far greater than this 0.2 to 30 magnetism range, which we have just considered. See note D, p. 624.

The most important point is that the difference in conditions between methods (1) and (5) ought naturally to lead to just this kind of difference in the ratio of hardness to magnetism. Step 1 of the transformation, from dense gamma to bulky beta, involving great expansion, tends to set up com-

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earliest period for which the data are given, at or before 3 hours at 500° and 650°, and at or before 6 hours at 400°, 550°, and 600°.

This contrast between the promptness of the hardness in reaching its maximum and the long-continued increase of the magnetism I interpret as meaning that the change from gamma to beta is relatively rapid, giving great hardness early, but that from beta to alpha slow. For ease of explanation I here use the terms of the beta hypothesis (§244). As this latter change progresses, the quantity of beta and hence the hardness are kept approximately constant by fresh transfers of gamma to beta about as fast as beta changes to alpha. The greater irregularity of the hardness than of the magnetism curves tallies with this conception, the quantity of beta depending on two independent variables:—the rapidity with which the gamma particles change to beta, and that with which these particles in turn change to alpha. Thus the quantity of beta is in flux, whereas that of alpha is constantly increasing.

The decrease of the ratio of beta to alpha, of hardness to magnetism, as the holdings at 400° and at 500° are prolonged, is shown by the dotted lines in Fig. 31, VI. The teaching is just what we should expect, not first a generation of beta without any alpha and then a generation of alpha from that beta, but a much more rapid generation of beta than of alpha, part only of the iron which has become beta having at any given instant changed further into alpha. Further, a progressive decrease of this ratio as the exposure to given temperature is prolonged, a natural result of the progressive depletion of the supply of gamma to replace the beta which is now changing into alpha. If this process could be pushed to completion it would, after exhausting the gamma, change the last of the beta into alpha, so that the ratio of beta to alpha would fall to zero, though there would be a slight hardening of the final alpha by the precipitation of cementite.

Additional evidence that the hardening, resulting from the gamma-beta change, is wholly independent of the increase of magnetism, resulting from the change of that beta into alpha, is given by the contrast at 400° and at 650° between the relatively rapid and great increase in hardness and the trifling increase in magnetism. 650° is evidently above *Ae*1, so that, though gamma changes readily to beta, beta does not change to alpha. 400° is evidently high enough to permit a relatively rapid change from gamma to beta, but not to allow that from beta to alpha. 550° and 600° lie close to the bottom of the transformation range in equilibrium, so that slight errors in temperature would affect the results greatly. Even the lag in taking the heating curves did not raise *Ac*1 for this material above "about 650" (Hadfield and Hopkinson, pp. 112–114). Additional sources of discrepancy may lie in differences of composition, with corresponding differences in the temperature of the transformation range.

[1] *Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, p. 112.

pressive stresses and thus to arrest itself. Step 2 does not. When, as in quenching carbon steel, but little time, and that at low temperature, is available for the transformation, this mechanical obstruction to step 1 is extremely effective, but there is no corresponding mechanical obstruction to step 2. Hence the ratio of the number of particles which having taken step 1 are able to take step 2 should be much greater here than when, as in method (5), the high temperature, say 650°, with its great mobility, and the long time available lessen greatly the effectiveness of the mechanical opposition to step 1. Clearly the great sluggishness of the transformation in method (5) represents chemical resistance. This case differs from that of quenching in that it does not offer to step 1 a mechanical impediment from which step 2 is free. Hence we may expect the ratio of beta to alpha to be much greater here than after the quenching of carbon steel, and this is exactly what we do find, much greater hardness for given magnetism after method (5) than after quenching carbon steel.

247. The amorphous theory[1] is that the interstratal movements, caused by the dilatation which accompanies the gamma-alpha or austenite-pearlite transformation, break up the crystalline condition of much of the iron and make it amorphous, at a temperature so low that this amorphous iron does not recrystallize, but is thus trapped in the hard amorphous state.

This theory is based on two undisputed facts: first, that in rapid cooling the transformation from gamma to alpha, whether it passes through beta or not, occurs at a relatively low temperature, and second, that amorphous iron is much harder than crystalline iron. The lowering of the temperature at which the transformation occurs by rapid cooling was proved by Osmond,[2] and is a phenomenon of undercooling or lag. It is in part explained by the fact that the transformation itself is a time-consuming reaction, which, setting in during a rapid fall of temperature, is naturally protracted till the temperature has sunk far below that at which it is due, *A*1.

[1] Compare Beilby, and Rosenhain, *Engineering*, May 15, 1914, pp. 661–2; *Journ. Iron and Steel Instit.*, 1914, No. 1, vol. 89, pp. 178, 180; Edwards and Carpenter, and McCance, *idem*, pp. 138 and 192. Edwards and Carpenter might be understood to mean that the reason why rapidly cooled steel is hard is that the rapid cooling has suppressed the transformation, and that this suppression in and by itself implies the existence of such interstratal movement as to generate a very great quantity of amorphous iron through twinning. There would be a certain confusion here, because a transformation which has not occurred is non-existent, and a non-existent thing cannot have any positive effect like that here ascribed to it. And it is not when the transformation is suppressed that the steel is hardened, but when it is allowed to go as far as the martensite state. If, as in the rapid cooling of steel rich in manganese or nickel, the transformation is wholly suppressed and the austenite state thus preserved, the steel is not hardened. Therefore, it is not the suppression of the transformation but its going as far as the martensite stage that causes hardening.

[2] "Transformations du fer et du carbone," Paris, 1888, p. 36. In an air cooling the temperature at which the cooling was arrested by the recalescence, or *Ar*1, was 29° lower than in a gentle cooling in his tubular furnace. It indeed remains to be proved thoroughly that the temperature of so much of the transformation as occurs in a truly rapid cooling is in fact low enough to prevent recrystallization. Certain it is that the sides of cracks which form in hardening steel are black, which goes to show that they cracked open at a fairly high temperature, at which recrystallization is not necessarily excluded. Of course the time available for such recrystallization is extremely brief.

Rosenhain[1] conceives the amorphousness as arising in a radically different way. Starting with the reasonable premise that every grain must needs be encased in a thin amorphous envelope (§514), he infers that rapid cooling, by leaving but an insignificant time for the growth of the grains of alpha iron which result from so much of the transformation as occurs, leads to extremely minute grain size and thus to extremely numerous amorphous envelopes, and thus in fine to a very great increase in the proportion of amorphous metal. The retarding effect of carbon, which has to be admitted whatever our explanation of hardening, he conceives to represent the time needed by the nascent alpha ferrite to expel its carbon to the surrounding untransformed austenite. Because carbon, though abundantly soluble in austenite, is insoluble in ferrite, we may imagine that the transformation of any given particle from austenite to ferrite cannot occur till it has succeeded in expelling its carbon to the untransforming austenite which surrounds it.

These two causes of amorphousness should, I believe, be taken jointly, because of the great increase which plastic deformation causes in the hardening effect of grain fineness (§489).

248. The greater hardness of the amorphous state is familiarly shown by the increase of hardness which wire-drawing and all other forms of overstrain cause in the metals in general, an increase which is now most reasonably referred to the amorphization which the mechanical distortion of the overstrain causes (§§506 *et seq.*, 628, 642). It is familiar also in the case of the silicates, which are harder when caught in the vitreous amorphous state, as when a slag is solidified too rapidly to crystallize, by pouring it while molten into water, than when they are allowed to crystallize, as when the same slag is allowed to solidify slowly, when it devitrifies, that is, crystallizes and thereby becomes softer.

249. What can Cause Amorphism?—Starting from these facts, (1) that rapid cooling lowers the transformation to a temperature at which iron if made amorphous could not recrystallize, so that amorphization if it occurred would persist, and (2) that amorphous iron is hard, the amorphous theory holds that the hardening caused by rapid cooling represents amorphizing. In short, it assumes amorphizing.

But how is this amorphizing to occur? It might be set up either mechanically or crystallographically. Mechanical distortion in the cold, as in wire-drawing, forcibly removes the constituent particles of the various crystal units so far from their relative positions, which are an essential part of their continuing to be crystalline units, that they remain unable to resume those relative positions, and in doing this it dis-crystallizes them. If this occurs at a high temperature at which the mobility is great enough to permit the dis-crystallized particles to remarshall themselves in crystalline order, though not necessarily in the same old groupings, they do this at once. If not, they are forced to remain dis-crystallized, that is, amorphous. Thus

[1] "Introduction to Physical Metallurgy," 1914, p. 257; and *Trans. Faraday Soc.*, 1915, vol. 10, p. 286.

the actual distortion in mechanical polishing and wire-drawing transfers much metal to the amorphous state (Chapter 20).

This form of amorphizing calls for actual translation of the particles of each amorphized crystal unit relatively to its fellows in that same unit. A simple expansion in which all join uniformly could no more amorphize them than the daily expansion and contraction of the transparent crystals in a mineralogical cabinet, as the room temperature rises and falls, amorphizes them. Such translation is caused not by pressure as such but only by distortion of form beyond the elastic limit as a result of unbalanced pressure. In rapid cooling the various concentric layers cool and hence contract at different rates, aeoliotachically, and this differential or aeoliotachic contraction might cause such distortion of the individual crystal units as to amorphize them.

The difficulties in imagining such mechanical distortion in the rapid cooling of a heterogeneous mass evaporate on examination. At first we say that if a sphere cools uniformly through the uniform escape of heat from every unit of its surface, and does not crack, there is indeed great internal tension and pressure because of the more rapid cooling and contraction of the surface than of the interior, but not translation of any particle beyond the centripetal movement due to the contraction. A nest of concentric compressible spherical layers subjected to hydraulic pressure would simply move centripetally by the amount of the contraction induced by the pressure, but would not slide over each other. But if we conceive a single sphere made up of certain incompressible particles imbedded in a compressible matrix, and assume that we compress it by hydraulic pressure, then because the surfaces of the incompressible particles do not change during such centripetal contraction while the surfaces of the compressible matrix at its contact with those incompressible particles do change, there must be relative movement between the former and the latter during this centripetal movement of the whole sphere.

In the same way with our sphere of austenite at the moment of quenching, as the transformation starts it generates certain alpha particles different in compressibility from the gamma particles; and, because of this difference, relative movement occurs at the surfaces of contact of the more compressible with the less compressible, and this relative movement may well amorphize the particles in which it occurs.

But the case is even more favorable. The transformation from gamma to alpha causes great expansion, as has long been inferred from the cracking of cylindrical steel objects in rapid cooling and has been shown conclusively by direct measurement.[1] Conceive now certain alpha particles suddenly generated within a matrix of untransformed gamma iron and expanding as they are generated. This very act of expansion implies an increase of surface, and

[1] Charpy and Grenet, *Bulletin Soc. d'Encouragement*, 1903, 1, p. 462; Benedicks, *Journ. Iron and Steel Inst.*, 1914, No, 1, vol. 89, p. 407; Charpy and Cornu-Thenard, *Journ. Iron and Steel Inst.*, 1915, No. I, vol. 91, p. 276. See Fig. 31, I, p. 179.

hence movement of that surface relatively to the surfaces of gamma with which the new-born alpha is in contact, because that gamma is not undergoing any corresponding expansion. True the whole mass is contracting because of the rapid cooling. But this simply means that the expansion which the gamma-alpha change implies is lessening *pro tanto* the contraction which the nascent alpha is undergoing, so that during the instant of its birth the surfaces of each alpha particle, because it is contracting less, must needs slide over the surfaces of the untransformed gamma with which it is in contact, because this is contracting more. Thus actual relative translation might well occur, and if it occurred it might generate amorphous iron, and if it occurred at a temperature too low for this amorphous iron to recrystallize the amorphousness would persist, and if amorphous iron persisted it would be hard. Hence rapid cooling might harden through causing amorphous iron, and hence the amorphous theory is tenable qualitatively thus far.

250. Crystallographic amorphizing, we may suppose, might arise from the occurrence of the transformation at a temperature so low that, on the breaking up of the gamma crystallization by the transformation itself, the resultant alpha particles lack the mobility to group themselves alpha-wise, and hence remain amorphous. This form of amorphization does not require any mechanical distortion or other form of relative translation of the various members of the individual crystal units. But in this case they are more likely to remain as pseudomorphs than to amorphize.

251. Is amorphization competent quantitatively to explain the degree of hardening caused by rapid cooling, and to explain the hardening which accompanies martensitization under the four other conditions already enumerated?

The hardening caused by rapid cooling is of an order of magnitude only moderately greater than that caused by the amorphizing of wire-drawing. That it is on the whole decidedly greater is readily understood from its occurring at an enormous number of surfaces throughout the mass, whereas the maximum amorphization which overstrain causes, for instance in wire-drawing to destruction, may well be concentrated along relatively few internal surfaces of low resistance. Destruction in wire-drawing will occur as soon as the supply of amorphizable crystalline metal has fallen so low along any one internal path of low resistance that mobile metal cannot be generated there fast enough by the ensuing fraction of the deformation to heal the ever-growing wounds.

That the martensitization which cold austenitic steel undergoes when subcooled in liquid air, or when overstrained by cold deformation, should harden the steel by amorphizing it is natural enough, because here the expansion

[1] Thus Goerens found that wire-drawing to destruction increased the hardness of steels of 0.11, 0.62, and 0.98 per cent. of carbon by 70, 76, and 27 per cent. respectively, whereas the martensitization of rapid cooling increases the hardness of such steels by about 54, 180, and 143 per cent. respectively (see Tables 25A, p. 466, and 34, p. 598) above that of the pearlitic or annealed state.

which accompanies the transformation should cause such a sliding of the nascent alpha surfaces past the remanent gamma surfaces as we have imagined in the case of rapid cooling, and because here the temperature is too low to permit the iron thus amorphized to recrystallize. Hence in these cases amorphism might occur, and if it did, it would persist, and hence would explain the hardening.

252. Objections to the Amorphous Theory.—A natural one is that, if rapid cooling thus causes such extended amorphization as to explain the intense hardening of high carbon steel, it ought to harden pure iron and the other pure metals and alloys in general, yet it does not do so to any significant degree. A complete answer is that, though it causes aeoliotachic contraction and hence sets up stress, it does not cause in the metals the internal sliding necessary to amorphization which martensitization causes. Here we should remember that, on the amorphous theory, it is not the rapid cooling or stress as such but the martensitization to which it leads that hardens carbon steel, the sliding at innumerable internal surfaces as each nascent expanding alpha article rubs against its unexpanding gamma neighbors. I have already pointed out that the particles which make up a homogeneous sphere need undergo absolutely no translation during rapid cooling, in spite of the tension and pressure set up. Without translation and consequent rubbing no amorphization is to be expected. And though rapid cooling certainly causes every non-spherical object to approach a little nearer to the spherical shape, yet the degree of approach is so very slight, and so nearly parallel and concentric in the various layers as not to cause any significant internal translation.

Hence the amorphous theory does not imply amorphization in the rapid cooling of pure metals or other homogeneous bodies, and hence the failure of such metals to harden on rapid cooling is consistent with that theory, and this present objection to the theory falls to the ground.

Two more serious objections remain. First, however competent the amorphous hypothesis may be to explain the hardening of carbon steel by quenching, it seems wholly incompetent to explain hardening by either method (2) or method (5). The great hardness given by method (2), Fig. 31, III, rising even to 578 or that of steel of 0.66 per cent. of carbon when quenched and untempered, is given statically and apparently at a high temperature, and is apparently not affected materially by any length of exposure to temperatures even as high as 800°. In method (5) the very material hardness, rising nearly to 450, is not lessened even by a 600-hour holding at 500° or by a 72-hour holding at 600° (Fig. 31, V). In such long high heating any metal made amorphous by slip certainly should recrystallize, and the fine alpha crystals of Rosenhain with their abundant amorphous envelopes ought to grow rapidly and thus reduce the number and hence the total quantity of these envelopes proportionally.

Second, in the first part of a subsequent heating or tempering the behavior of steel hardened by quenching differs radically from that of iron hardened

by being made in part amorphous by plastic deformation, the former beginning to lose hardness with the very first reheating, the latter gaining in hardness and elastic limit (Fig. 78 and §631).

On the other hand, the amorphous theory receives a certain support from the fact that the hardness of certain copper aluminum alloys is about doubled, with simultaneous martensitization, by quenching from above their transformation range (§608). This case needs much further study. The martensitic pattern need not imply that the immediate cause of the hardening is the same as in the hardening of carbon steel, because any sort of change occurring along the strongly marked octahedral cleavages of gamma iron would give suggestions of martensitization. Should it prove that the hardening of these copper alloys is to be explained, not by an intermediate allotropic state comparable with beta iron, but by solution or by amorphism or both, then we may have to admit that one or the other or both these causes may contribute to the hardening of steel by quenching. But the reasons why neither of these causes can explain hardening by methods (2) and (5) would retain their present force. Here we should remember that the hardness of these copper alloys is only doubled, whereas the hardness of hardened steel may be nine times that of alpha iron (Table 34, p. 598).

If we infer from the evidence already given that the hardening of iron by these two methods is referable to beta iron, we incline strongly to hold that this same beta iron plays a very important part in hardening carbon steel by quenching, yet this does not exclude contributory causes such as amorphism under these dynamic conditions. Indeed, the slightness of the loss of hardness in the first stages of reheating hardened carbon steel, as shown in Fig. 123, suggests that here the loss of part of the hardness due to the presence of beta iron is partly counteracted by a simultaneous gain of hardness representing the hardening induced by this same reheating in the part of the metal which has been made amorphous by the plastic deformation during the quenching.

253. Stress Theory.—Still a fourth explanation refers the hardening of steel by rapid cooling to the stress which certainly arises,[1] because of the aeoliotachic cooling and hence contraction. Stress no doubt contributes,[2] but it seems quite incompetent to explain the very great gain of hardness on rapid cooling, or the hardening caused by the martensitization of manganese and nickel steel without rapid cooling, first because pure iron and the other metals which do not martensitize on rapid cooling do not harden, though they must retain great residual stress, and second because objects the shape of which should lead to the least stress in fact harden the most intensely, such as very thin wire and the thin edge of a razor. Here the degree of aeoliotachicness is necessarily slight, yet the hardening is intense.[3]

[1] See Åckerman, *Journ. Iron and Steel Inst.*, 1879, No. II, p. 504.

[2] Thus unannealed glass is reported to be harder than the same glass when annealed, though so far as we know the only difference is in the release of stress in annealing.

[3] See the Author, *The Metallurgy of Steel*, 1890, p. 35.

But while macroscopic strain due to aeoliotachic contraction is incompetent to explain hardening, the microscopic strain due to martensitization, to the sudden expansive birth of alpha particles throughout the gamma mass, might be an important cause. The hardening effect of this microscopic strain would be very difficult to distinguish from the hardening caused by the amorphization proper.

254. Summary of the Discussion of the Cause of Hardening.—Apart from hardening by cold deformation, which is generally thought to act by amorphizing part of any ductile metal, there are five distinct methods, three of them dynamic and two static, three involving temperature changes and two at stationary temperature, of bringing iron to a hard state intermediate between the soft gamma and the softer alpha states. The hardness of mixtures of this hard state with one or both of the soft end states is so great as to indicate that the hardness of the hard state, if it could be completely isolated, would be some ten times that of alpha iron and far greater than that of gamma.

I infer that this intermediate state is not amorphous iron, first, because this hardening can be induced statically and thus without mechanism capable of causing amorphism; second, because it endures through long heating at temperatures as high as 800° (method (2)) at which amorphous iron would become crystalline and fine-grained iron would coarsen and thus reduce its amorphous envelopes to a relatively negligible quantity; and third, because the hardness, for instance of quenched carbon steel, decreases progressively on the slightest heating, whereas that of iron made amorphous by cold deformation increases on gentle heating, so that the hard constituent of such hardened steel differs strikingly from the amorphous hard constituent of cold-deformed iron.

Amorphism as the common cause of hardening by these five methods being thus eliminated, there remain the solution hypothesis that the intermediate hard state is a simple solution in varying proportions of the two end states in each other, the high-temperature gamma and the low-temperature alpha; and the beta hypothesis that it is a distinct beta state or states of iron, coördinate with the end states gamma and alpha. Gamma iron alone is dense, beta I and II alone are hard, and beta II and alpha alone are magnetic, both gamma and alpha being isolated easily, but beta thus far always occurring mixed with either gamma or alpha or more often both.

Between these I choose the beta hypothesis confidently, first, because the hardness of the hardened metal is far too great to represent a simple solution of the two soft end states in each other, especially in view of the slight increase of hardness which the dissolving of the glass hard cementite causes in gamma iron, and because this extreme hardness points to the presence of a far harder allotropic variety; second, because the sharpness of the inflection of the curve of the temperature coefficient of electric resistance, and the sharpness of the extinction of magnetism, at or close to $A2$, are so marked as to represent an allotropic change here rather than a mere progressive

change in the proportions in which alpha and gamma are dissolved in each other; and third, because, for given degree of hardness, the variations in magnetism are far too great to be explained by variations in the proportions of only two substances gamma and alpha, for such an admixture should, for given hardness, have only one or at most two relatively fixed degrees of magnetism, whereas the beta hypothesis of one hard mixed with one bulky and one or two magnetic substances might give almost any intermediate combination of hardness and magnetism. This third reason is reinforced by the fact that the variations in the ratio of hardness to magnetism with varying conditions of hardening are those which should be caused by the greater bulkiness of the beta and alpha than of the gamma state.

Nevertheless, in view of the hardening, slight though it be, of the copper aluminum alloys by quenching from above their transformation range, referable perhaps to the dissolving of the end constituents in each other and perhaps to amorphism caused by the quenching strains, it may be that, in the hardening of carbon steel by quenching, both solution and amorphism may contribute to the resultant hardening, that solution may in all the other hardening methods, and that amorphism may in all the dynamic methods.

The solution and the amorphous hypotheses, though each fitting some of the facts, yet fail by themselves to explain others. The beta hypothesis thus liberally applied seems to me to fit all the present and intricately related facts, many of them at first sight contradictory, with a degree of accuracy which justifies us in adopting it provisionally with some confidence for working purposes.

Because the solution of cementite in gamma iron hardens it so little, and because the hardness given by the static methods, which exclude amorphism, is nearly as great as the greatest given by quenching, I infer that the contribution of solution and amorphism to the hardening of steel is of minor importance, and the preservation of much iron in the beta state the dominant cause.

255. Steel of 0.22 Per Cent. of Carbon, Ordinate l–l^{VI} of Figs. 23 and 24.—The general course of events in the cooling and heating of this steel is like that of the steel of 0.40 per cent. of carbon, with the obvious differences that the freezing range $l^{I}l^{II}$ is somewhat higher, and that the generation of ferrite begins at a somewhat higher point, l^{III} instead of j^{III}.

As in the previous case, with the sinking of the temperature from l^{III} toward l^{V}, the residual austenite out of which the ferrite springs continuously approaches the composition S, 0.90 per cent. of carbon, temperature and carbon content sliding down GOS; and on cooling past the line PSK the austenite then remaining changes abruptly into pearlite, with recalescence. Hence the metal both now and when further cooled consists of a conglomerate of the ferrite which has formed in cooling from l^{III} to l^{V} mixed with the pearlite formed in passing l^{V}.

When the steel thus cooled is heated again, it passes through these same changes in the reverse order.

256. The genesis of the microstructure of this 0.22 carbon steel is indicated in a general way in Diagram III of Fig. 30, p. 163, in the rectangle bounded above by $A3$ and below by $A1$, for this steel. Note that early in the transformation, when the quantity of the white ferrite which has thus far been born of the black mother austenite is relatively small, the network structure is prominent, for the ferrite marks clearly the boundaries of the various austenite grains out of which it has migrated. As the temperature falls further, the quantity of ferrite is so great that the network structure, the relation of enclosing shell to enclosed kernel, is no longer prominent.

257. The transformation of steel of 0.22 per cent. of carbon is shown micrographically in Figs. P to S, Plate 9, which represent a series of pieces of this steel heated well above the transformation range so as to convert them into austenite, then cooled slowly, each to a given temperature below $A3$, that is, within the transformation range, in order to induce the generation of ferrite corresponding to that temperature, and then quenched in water so as to preserve the outline of the ferrite and austenite masses resulting from this generation. Fig. P shows the beginning of this process, at 825° or about 20° below $A3$, the formation of fine feathery lines of ferrite, outlining the austenite grains which have expelled this ferrite to their surfaces. Fig. Q shows the structure at a temperature 35° lower, about one-third way down from the top toward the bottom of the transformation range. The ferrite still outlines grains of austenite of a size corresponding to those of Fig. P, and so sets up an imperfect network system. But the network itself, instead of being continuous, is formed of chunks of ferrite which here look isolated, though they may well be united outside the plane of the section. The chunkiness of these masses reflects the work of surface tension in drawing each toward the spherical form, while their sharpness reflects the simultaneous effort of the crystalline force. We may suppose that these two are constantly striving for the mastery. Surface tension having drawn or balled a given mass well together, and thus nearly satisfied itself, the next particles to deposit obey the crystalline force and form sharp apexes on the rounded masses on which they are thrown down. Indeed, even a rudiment of an apex locally hastens precipitation, and thus opposes the effort of surface tension to round it.

Fig. R shows the condition reached at 760°, or about two-thirds way down from the top toward the bottom of the transformation range, and S that at 705°, near the bottom of that range and with correspondingly increasing quantities of ferrite. A specimen like that of Fig. S, if quenched from just above $A1$, would have its black ground mass martensite or more usually troostite, the austenite which existed just before quenching transforming as far as martensite or troostite even during a rapid cooling. If quenched from just below $A1$, or at any lower temperature, its black ground mass would be pearlite, too fine to show its banded structure under this magnification.

In Fig. R the network structure is greatly obscured, and in Fig. S it can hardly be traced. Here it is probable that the mass of ferrite so far

preponderates over that of the austenite that its surface tension dominates. Just as on first cooling past *A*3 the austenite grains, forming nearly the whole mass, could readily expel the small quantity of ferrite now existing to their boundaries, so the greater mass of ferrite which has come into existence at the bottom of this range is able to displace the residual austenite, and thus not only to conceal but actually to break down the pre-existing austenite grain system.

The structure at stage S of the ferrite itself, with its familiar etch-figuring, is shown in Fig. T.

Crystalline force is, of course, only a special directive manifestation of cohesion. But this manifestation has such strong characteristics of its own that it needs a specific name, quite as surface tension does, which quite in the same way, is only a manifestation of unbalanced cohesion. Hardness, tensile strength, and compressive strength are all manifestations of cohesion, but none the less they are given these special names for convenience.

258. Steel (?) of 0.00 Carbon, Ordinate AO, Fig. 23.—This metal freezes as a whole at a single point, 1,530°, instead of throughout a freezing range as in the previous cases. Clearly no differentiation or selection is possible in the freezing of a pure element.

On the beta theory the metal changes as a whole at *G* into beta ferrite, and this in turn into alpha ferrite at *M*, and this alpha ferrite persists unchanged till the cooling is complete. The change from beta to alpha, is probably a progressive one, spread out over a wide range of temperature, and *M* is to be regarded rather as the temperature of most rapid change than as one at which is concentrated the whole or even a large fraction of that change. There is much to indicate that the change at *G* also is a progressive one, spread out less than that at *M* (§228).

That this metal should freeze at a single point instead of throughout a range as in previous cases, that it should change as a whole at or near a single point into ferrite instead of throughout a long range which ends at the line PSK as in previous cases, and that it should lack the recalescence are but natural consequences of its completely lacking carbon. For the extension of solidification throughout a range, and of the transformation from austenite into ferrite (or cementite) plus pearlite throughout a wide lower range, were accompanied by a change in the distribution of the carbon, and as we shall see later, are due to that change. The case is strictly parallel to that of the solidification of our nitrate solutions. This is selective and progressive, extending through a range from the liquidus to the solidus. But when we pass to the extreme of the series, pure water, that solidifies unselectively and at single temperature, 0°. If any nitrate is present, its presence causes the freezing to be selective and progressive, concentrating the nitrate progressively into the residual mother-liquor. If the quantity of nitrate is zero, there is nothing to select.

Further, the recalescence is strictly a phenomenon of the change into

pearlite of so much of the austenite as has been brought by the transformation to the carbon content of 0.90 per cent.; and of course if there is no carbon no austenite can thus be brought to that carbon content.

This transformation is not readily shown micrographically, because, in the total absence of carbon and other retarding elements, even the most rapid quenching seems impotent to restrain the transformation far enough to leave any trace of martensitic structure. Instead the austenite structure, which can be developed by etching the metal when above A3, changes to that of ferrite during even the most rapid cooling.

The structure of this ferrite in nearly pure electrolytic iron is shown in Fig. A of Plate 3.

259. Steel of 1.45 Per Cent. Carbon. *Solidification.*—In cooling from the molten state this steel begins solidifying as austenite when its temperature reaches k^{I} (Figs. 23 and 24); its solidification becomes complete when its temperature reaches k^{II}; and the solid white hot austenite cools without further change till it approaches k^{III} on *ES*, in every respect paralleling the behavior of the steel of 0.40 per cent. of carbon.

Transformation.—From this point on the parallelism is incomplete in one respect, in that it is not ferrite but cementite, Fe_3C, that the individual yellow-hot grains of austenite now begin generating and ejecting as they enter the transformation range. But they treat this cementite quite as the 0.40 per cent. carbon steel treated the ferrite which it generated. Each austenite grain ejects to its outer surface part of the cementite to which it gives birth, thus enclosing itself as an austenite kernel within a cementite shell, and it masses along its own cleavage planes much of the remainder of this cementite. As the temperature descends further, this mother austenite continues generating cementite, so that the cell walls continually thicken at the expense of their kernels.

Because this cementite contains 6.67 per cent. of carbon, the mother austenite, which initially contains only 1.45 per cent. of carbon, in the act of generating this cementite of course impoverishes itself in carbon. This impoverishment takes place at such a rate that the carbon content of the residual austenite at each temperature is measured by the abscissa of the curve ES at that temperature; and in particular that, when the temperature has fallen to k^{IV}, the carbon content of the residual austenite has fallen to S, or 0.90 per cent., and the austenite itself has thus reached the eutectoid or hardenite carbon content.

On cooling past k^{IV} this residual austenite changes as in the previous cases to pearlite. In further cooling no change needing our attention occurs, and the cold steel consists of this pearlite plus the cementite which has been generated between k^{III} and k^{IV}.

If the steel is now re-heated it passes through these same changes in the reverse order; and as in previous cases, the change back from pearlite and cementite into austenite occurs at a temperature a little higher than that at which the opposite change occurred in cooling down.

This pro-eutectoid transformation thus parallels the pro-eutectoid transformation of hypo-eutectoid steel, and the pro-eutectic solidification of hyper-eutectic nitrate solution and of hyper-eutectic cast iron, substituting the birth of pro-eutectoid cementite for that of pro-eutectoid ferrite, primary nitrate, and primary cementite respectively.

260. The Quantitative progress of solidification and transformation is shown in Diagram III of Fig. 24. There is the progressive increase in the quantity of solid black austenite from k^{I} on the liquidus to k^{II} on the solidus, the generation of pro-eutectoid cementite from k^{III} on the liquidoid SE to k^{IV} at $A1$ or PSK; and on passing $A1$ the transformation of the residual austenite, now enriched up to the eutectoid carbon content 0.90, into pearlite, which persists unchanged down to the atmospheric temperature.

261. A sketch of this transformation is given from k^{III} to SK in Diagram IX of Fig. 30, §220, representing this steel. Note that, on cooling below k^{III}, that is on passing below the line SE of Fig. 23, a network of cementite starts to form, and that it increases in thickness as the temperature falls toward SK or $A3$–1–2, 725°. On passing this point the black austenite as in all previous cases changes sharply into striped pearlite, with the familiar recalescence, and the simultaneous loss of the hardening power.

262. The Progress of the Transformation Illustrated Micrographically.— Figs. A to H, Plate 9, show the progress of the formation of cementite within the solid red-hot metal as it cools from k^{III} to k^{IV}, and Figs. I to N, Plate 9, show its progressive re-absorption as the temperature rises from k^{IV} to k^{III}. Note the formation of the cementite cell walls in Fig. B, the thickening of the walls and the shooting out of spines within the cells in Figs. C to F.

These specimens A to H were prepared as those shown by Figs. P to S of Plate 9 were, and as explained in §257, substantially by heating to above the transformation range so as to convert the whole into austenite; cooling the various specimens each to a predetermined temperature within or below that range, so as to induce the precipitation of the quantity of pro-eutectic cementite due at that temperature; and quenching in water so as to fix the structure thus reached. The dark ground mass of those quenched from below A1 of course consists of pearlite, whereas that of those quenched from above A1 consists of the products of the transformation, during quenching, of the austenite which existed at the moment of quenching. In these specific cases the contrast between the dark ground mass and the white pro-eutectoid cementite was exaggerated by reheating the quenched specimens to 300° before polishing them, so as to convert the martensite into the black-etching troosite, but without material alteration of the outlines of these masses. The micrographs show little difference between the troostitic ground mass of those quenched from above A1, Figs. C to F, and the pearlitic ground mass of those cooled slowly to below A1, Figs. G and H, for of course the pearlitic structure cannot be resolved with the magnification here used.

In like manner Figs. I to N show the reverse progress in heating up

from k^{IV} to k^{III}, first the change from the lighter pearlite ground mass to the darker austenite, transformed back into troostite in the quenching, in Fig. K; then the thinning out of the spines of cementite within the cells, Fig. L; later the dissolving away of the cementite shell-walls, Fig. M, and last their almost complete obliteration, Fig. N.[1]

Note that the pro-eutectoid cementite, like the pro-eutectoid ferrite, is distributed by the mother austenite grains in part as a network surrounding those grains, and in part in their parallel octahedral cleavages. In Fig. 30 this resemblance is easily traced. The cementite needles are usually much straighter and thinner than the ferrite masses, a difference which may be interpreted as meaning that the mother austenite has far better control over the distribution of this small quantity of cementite than over the much larger quantity of ferrite in a hypo-eutectoid steel equally distant from the eutectoid carbon content. In like manner, the distribution of the small quantity of ferrite in steels but slightly hypo-eutectoid is usually much more regular than that of the greater quantity present in very low-carbon steels, the great majority of ferrite here resisting the sway of the minority of austenite.

263. The quantity of pro-eutectoid cementite looks to the eye much greater than that which is due, in Figs. E and F, Plate 9. Part of the apparent excess may represent a labile precipitation of cementite from a supersaturated solution, which would include part or even the whole of the cementite which ought to be present as pearlitic cementite. But even if the whole of the carbon present, 1.45 per cent., were thus precipitated it would form only 21.75 per cent. of cementite, whereas in places the apparent quantity is much greater than this. Hence part of the apparent excess must be due to an optical illusion,[2] which makes the area of these fine white lines seem greater than it is, as often happens when the magnification is small.

264. Incipient Fusion Shown Micrographically.—Having traced thus micrographically the changes which occur during transformation, we see in Fig. A, Plate 10 the change which occurs on rising past the solidus at k^{II} in a hyper-eutectoid steel of about this composition. This micrograph we owe to the skill of Prof. K. W. Zimmerschied.[3]

This specimen was heated slightly above the solidus, and then quenched in cold water to preserve the *status quo* as far as might be, though even such a cooling cannot prevent the change of part of the austenite into the characteristic zigzag martensite needles, lying in those octahedral cleavages

[1] These micrographs represent a series of specimens of the same steel which were heated to the temperatures, shown, held 30 minutes, and quenched, after a preliminary high heating to develop a coarse network structure, followed by slow cooling.

[2] See Howe and Levy, "The Life History of Pro-eutectoid Cementite." (*Internat. Congress for Testing Materials*, 1912, II, 4 and discussion.) On larger magnification the apparent quantity of pro-eutectoid cementite in this specimen becomes very much smaller than that here shown. Beyond this, the cementite is not evenly distributed, quite as the quartz in a granite may not be.

[3] *Trans. Amer. Inst. Mining Eng.*, 1913, vol. 47, Fig. 15, p. 653.

of the austenite into which, in slow cooling, the pro-eutectoid ferrite of low-carbon and cementite of high-carbon steel are coerced.

In order to interpret this micrograph, let us remember that, in the preceeding cooling through the transformation range, the then austenite has thus assembled in the grain boundaries much of the pro-eutectoid cementite to which it gave birth, and that these boundary regions were thereby made richer in carbon than the interiors of the grains. In again heating through the transformation range this cementite is progressively re-absorbed by the mother austenite, and this re-absorption is due to complete itself in reaching the liquidoid SE, Fig. 23, p. 130, and this concentration of carbon there tends to efface itself by diffusion. But this effacement, however rapid at first, yet takes time to complete itself, indeed probably completes itself fully only asymptotically. When the temperature in rising reaches the solidus, this diffusion may yet be incomplete, with the consequence that the solidus for the carbon-richer grain-boundary regions corresponds to a lower temperature than that which corresponds to the solidus of the carbon-poorer interiors; or in short the richer grain-boundary region is more fusible than the poorer interior. Hence the beginning of fusion at the grain boundaries, and particularly at what might be called the "cross roads."

We might, indeed, expect the subsequent solidification to efface the traces of this slight re-melting, as it would if it proceeded exactly as the initial solidification of the mass does. We do not find the parts last solidified during the initial solidification thus differing from the neighboring parts which solidified slightly earlier. The reason for this is that, this initial process of solidification being a continuous one, there is complete continuity of the crystallizing conditions which determine the exact structure with which the metal solidifies. But when this present slight re-fusion is replaced by re-solidification, the crystallizing conditions could hardly be expected to be identical with those of the initial solidification, and hence the structure of the re-solidified part could hardly be expected to be identical with that of the unmelted part on which it re-solidifies. Indeed in this case the steel, while thus very slightly re-melted, was quenched in water, so that the conditions of the re-solidification differed radically from those of the initial solidification. Hence one reason for the difference of shade between the strips which have melted and the mass of the grains.

Even were the differentiation caused in the pro-eutectoid transformation later effaced by diffusion in a long stay above the transformation range, there would still be local differentiation on rising slightly above the liquidus; for, just as the solidification of the molten mass is a differentiation, the carbon content of the molten represented by the liquidus AB and that of the depositing solid by the solidus AE, so in re-melting this same differentiation sets in. The molten again approaches the carbon content represented by the liquidus, and the shore layers against which it abuts approach the carbon content represented by the solidus, carbon passing from shore layers into molten to bring about this differentiation. But that which determines

that this differentiation shall follow this particular path, that the region in which fusion starts shall be the grain boundaries rather than somewhere else, may be taken to be the residual enrichment of those boundaries in carbon, persisting in spite of the opportunity which there has been for diffusion.

265. Incipient Fusion of a Non-eutectiferous Steel may Give Rise to a Eutectic.—The specimen shown in Fig. B, Plate 10, is like that of Fig. A, with the exception that the re-melting has been carried a little further. This too is due to Prof. Zimmerschied.[1] As in Fig. A, so here the slightly re-melted steel was next quenched in cold water. Here the eutectic is shown in typical concave or spandril masses, female to the male crystals of solid austenite which it bathes (§§174 and 186). Only in certain of the re-melted spandrils can the eutectic structure be seen. The remaining ones, though possibly eutectic too fine to be resolved by this magnification, may not be eutectic at all.

To judge from Fig. A of Plate 10, this steel before this slight re-melting was wholly free from eutectic. That it should form eutectic on this re-melting is natural, in view of the reasoning in §207. Turning to Fig. 29, p. 151, this steel, just as it starts to melt, is represented by point *G*. When thus slightly re-melted it splits up, under conditions of equilibrium, into a solid of carbon content g^{I} and a molten of composition g^{II}. Suppose that this molten were sucked out of its present place and allowed to solidify in a magnesia mould. Because it lies so far at the right of *E* it must, when thus solidified, be composed almost wholly of eutectic. If, instead of being thus sucked out, it remained *in situ* and solidified so extremely slowly that its excess of carbon over *G* had time to diffuse back into the solid, then when the last solidifying layers reached *G* the quantity of molten would have fallen to zero. But just as in §207, so here any incompleteness of diffusion widens the eutectiferous range, and it might easily widen it enough to cause much of this molten steel to re-solidify as eutectic. In fact, the quenching of this mass in water prevented diffusion almost completely, so that the part which had been re-melted just before the quenching, in spite of being present within the enclosing solid steel, in effect behaved during that quenching nearly as if it were solidifying in an inert magnesia mould, and therefore re-solidified chiefly as eutectic.

[1] *Loc. cit.*, Fig. 17, p. 655.

CHAPTER 10

INTRODUCTION TO THE CARBON-IRON DIAGRAM (CONTINUED)

(C) The Transformations in Cast Iron; Summary of the Carbon-iron Diagram

266. Summary.—The transformation in cast iron is essentially that of the austenite which it contains, and because this austenite is identical in composition, at the moment of solidification, with a steel of 1.70 per cent. of carbon, the transformation in cast iron is identical with that of such a steel.

This austenite exists in two forms: the eutectic austenite which is the filling of the eutectic honeycomb, present in all cast irons; and primary austenite, or that which solidifies selectively from molten hypo-eutectic cast irons, those with less than 4.30 per cent. of carbon, during the first or pro-eutectic stage of the solidification, from its beginning down to the eutectic solidification temperature of about 1,135°.

267. The Transformations in Eutectic Cast Iron.—Taking up the case of our eutectic cast iron where we left it in §172, when it had cooled to below the eutectic freezing-point and had solidified, we may now consider its behavior in cooling down through the transformation range, Fig. 23, p. 130.

We saw in §259 that, in the cooling of steel of 1.45 per cent. of carbon, the austenite of which it consists when it first solidifies begins to be transformed as soon as its temperature in cooling reaches the transformation line *SE*, Fig. 23, by the gradual birth or generation of pro-eutectoid cementite within it, a generation which continues, with progressive impoverishment in carbon of the residual austenite, till the temperature descends to 725° or *A*1 and the carbon content to 0.90. But the same thing is true of the eutectic austenite, which is represented by the black austenoid filling of the honeycomb eutectic formed in the solidification of this 4.30 per cent. cast iron, as seen for instance in Fig. B of Plate 12. Of course because this austenite in our present cast iron initially contains some 1.70 per cent. of carbon, it is in the very act of birth on the line SE, so that its transformation begins as soon as the temperature starts to descend below the freezing-point or 1,135°. In other words, as the cast iron, which in the act of solidifying was split up into the eutectic, a conglomerate of cementite plus 1.70 per cent. carbon austenite, begins cooling below that freezing-point, though this cementite remains intact, this austenite is progressively impoverished in carbon just as if it were not accompanied by that initial cementite. We have already seen that this eutectic austenite is identical in composition with a 1.70 per cent. carbon steel (§81), and we have called the special specific conglomerate of pearlite with pro-eutectoid cementite into which it transforms in slow cooling "1.70C austenoid" (§80).

The rate of that impoverishment is such that the carbon content of the residual austenite at any temperature is measured by the abscissa of the line *SE* for that temperature, and becomes 0.90 per cent., *i.e.*, has the eutectoid composition, when the temperature has sunk to *A*1.

On further cooling this austenite of eutectoid carbon content, "hardenite" (§211), changes into pearlite exactly as in the case of the several steels which we studied in §§220 to 263. Thus, though the path of the cooling cast iron as a whole is represented by points successively lower on the ordinate Bn^{I} the iso-carb of 4.30 per cent., yet of the two constituents of which it is composed the cementite is represented persistently by the ordinate *CKL*, of 6.67 per cent. carbon, which is the iso-carb of cementite, and the austenite at all temperatures down to *A*1 by the line *SE*; and at all lower temperatures by (1) the ordinate *CKL*, which represents both the initial cementite and that which forms within the pearlite on crossing *PSK* (*Ar*1), and (2) the ordinate *PN*, which represents approximately the ferrite of that pearlite. It is in this sense that this cast iron while in Region 5 consists of cementite plus austenite, and when in Region 8 of cementite plus pearlite.

268. The quantitative progress of the transformation is shown in Diagram V of Fig. 24 by the progressive increase of the quantity of pro-eutectoid cementite, marked "*W* from *X*," and the corresponding decrease of the quantity of the eutectic austenite *X* out of which it springs, from *B* to *A*1, and the change of this austenite into pearlite on cooling past *A*1. These various constituents are shown here hatched with the coarse eutectic hatching, to represent the fact that the eutectic maintains its main outline, and on small magnification gives little outward sign of undergoing these transformations.

269. The transformation is illustrated micrographically by Figs. A and B of Plate 12, of which the former shows a specimen of cast iron, actually of spiegeleisen, quenched from 1,050° in order to retain the structure which existed there, and the latter shows the same iron in its initial slowly cooled state.

In Fig. A the smooth white masses are the eutectic cementite, and the rougher dark ones are the eutectic austenite, which here seems to have been preserved nearly intact, with but little change forward toward the state of martensite, thanks to the obstructive action of the manganese present.

In Fig. B the white masses are as before the eutectic cementite, and the dark ones are the pearlite into which the austenite has transformed in cooling past *Ar*1.

Thus far I have not succeeded in showing the pro-eutectoid cementite which certainly should form during the slow cooling of the eutectic austenite, as it does in the slow cooling of any hyper-eutectoid steel and in the primary austenite of cast iron[1] (§260). Indeed, it is only by taking advan-

[1] It is not certain that the precipitation of pro-eutectoid cementite was abundant in this specific specimen, which is very rich in manganese. In such an iron the line SE may, for all we know, be so much steeper than in Fig. 23 that only a small quantity of pro-eutectoid cementite forms.

tage of the extremely coarse structure of the eutectic in this specimen of spiegeleisen and by magnifying even this very greatly, × 1,200, that it has been possible to show clearly the pearlite into which this eutectic austenite transforms. The subdivision of the eutectic austenite into distinct grains is seen in Fig. A, and it is strongly suggested in Fig. B by the rather sharp changes in structure and orientation of the pearlite, along what were presumably the austenite grain boundaries before it transformed into pearlite.

This alloy is actually hyper-eutectic. Nevertheless, the transformation which occurs in the austenite of its eutectic serves to illustrate that which occurs in a strictly eutectic cast iron, and would no doubt be disclosed in such an iron could we find one with a coarse enough structure.

270. The Transformations in 3 Per Cent. Carbon Hypo-eutectic Cast Iron.—This case resembles the last, with the difference that there is not only eutectic austenite but also primary austenite to transform into austenoid, simultaneously and in like manner. That is to say, this generation of pro-eutectoid cementite occurs not alone in the austenite which forms the black bands of the eutectic, but also in the primary austenite which forms the dendritic black masses, *e.g.*, in Figs. A, B, and C of Plate 6. Hence at all temperatures between the solidus, 1,135°, m^{II} of Fig. 23, at which its solidification completes itself, and $A1$ or m^{III} at which its transformation completes itself, this cast iron is a conglomerate of (1) primary austenite with (2) eutectic austenite, both containing pro-eutectoid cementite generated in this Region $5A$, and (3) the eutectic cementite unchanged. On cooling past $Ar3$–2–1 or PSK, about 725°, both the primary and the eutectic austenite, which have been impoverished by this genesis within it of pro-eutectoid cementite, down to the eutectoid or 0.90 per cent. carbon content, and have thus become "hardenite," transform as in all like cases into pearlite, with the usual recalescence, which is of course greatly obscured by the presence of the unrecalescing eutectic cementite. In Region 8, *i.e.*, at all temperatures below $A1$, it is a like conglomerate with its austenite transformed into pearlite, that is, it is austenoid.

271. The quantitative progress of the transformation is shown in Diagram IV of Fig. 24, §169. It is quite like that of the eutectic cast iron considered in §267, with the exception that there is generation of pro-eutectoid cementite not only from the eutectic austenite but also from the primary austenite, so that between m^{II} on the solidus and $A1$ there are two areas of progressively increasing pro-eutectoid cementite, marked "W from U" and "W from X" respectively.

272. The transformation of hypo-eutectic cast iron is illustrated micrographically by the series of figures on Plate 11. Figs. A to F inclusive represent pieces of one and the same chilled cast iron, which were heated to 1,050° side by side, then cooled to the various temperatures indicated, and then drawn severally and quenched in water to preserve the structure

reached at these various stages of the transformation.[1] Fig. C of Plate 12 represents a similar cast iron, the "washed metal," of Fig. B of Plate 6, which has been cooled slowly to the room temperature.

In all these figures the lighter smooth-etching mass is the eutectic cementite, the small dark islands within it represent the eutectic austenite, and the large dark areas represent the primary austenite. The bold convex outline of these last areas, with their progressive structural change, with falling quenching temperature, from coarse to fine martensite, to troostite, and thence to pearlite, suffice to identify them as representing what was initially austenite. The smoothness of the white masses, their spandrel shape, and to a certain extent their containing round cells filled with dark austenoid, identify them as the eutectic cementite. But because of the undercooling caused by the rapidity with which this iron was forced to solidify, the quantity of eutectic austenoid is abnormally small, and indeed in many places there is none, the eutectiform mass consisting of cementite only.[2] This undercooling is explained briefly in §274.

Note the contrast between the constancy of the structure of the cementite from end to end of the series of Plate 11, and the progressive changes in the darker constituent. In Fig. A it has the coarse martensitic structure which is so readily had by quenching high-carbon austenite from a high temperatures. The specimen of Fig. B, quenched from 765°, has the finer and less well marked martensite characteristic of these lower quenching temperatures; while in Fig. C, quenched from 728°, the speed of cooling through the transformation range was so slow, and for other reasons[3] the obstruction of the transformation was so much less complete, that the metal has transformed beyond the stage of martensite to that of the black-etching troostite. In Figs. D, E, and F of Plate 11, and C of Plate 12, representing specimens cooled slowly to below Ar3–2–1, the dark areas are pearlitic, though, as often happens in white cast irons, the presence of so much non-pearlitic cementite has led to the divorcing of the pearlite from its typical lamellar structure as explained in §273. Instead of this we find a sea of ferrite with some rounded islands of cementite, much of what should be the pearlitic cementite evidently having coalesced with the great masses of the eutectic cementite. This divorce of the pearlite resulting from the transformation of the primary austenite does not always occur in white cast irons.

[1] These specimens were held for 45 minutes at 1,050°; then cooled rather rapidly toward 765° by transferring them to a bath of molten lead at that temperature, and thence cooled moderately slowly in the molten lead to their various quenching temperatures. About 5 minutes elapsed between the entry into the lead bath and reaching 765°. The purpose of cooling at this intermediate rate was to give time enough for the precipitated pro-eutectoid cementite to coalesce into masses visible under this magnification, yet not enough for this cementite to ball up under surface tension, or to obey its tendency to join the great masses of eutectic cementite which lie so near.

[2] Howe and Levy, reply to discussion of their paper, "Notes on Divorcing Annealing and Other Features of Structural Coalescence," *Proc. Cleveland Inst. Eng.*, 1913–1914, p. 218.

[3] The Author, "Why Does Lag Increase with the Temperature from Which Cooling Starts?" *Trans. Amer. Inst. Mining Eng.*, 1913, vol. 45, p. 516.

Thus undivorced lamellar pearlite is seen in Fig. B of Plate 13, which shows the face of a chilled cast-iron car wheel.

The progressive generation of pro-eutectoid cementite in the primary austenite as it cools is here prominent. Note that, though the primary austenitic masses of Fig. A of Plate 11 are free from cementite, those of the remaining figures contain a very considerable quantity. It is recognized as the long thin white needles, which in Figs. B, C, and D are strictly parallel to two or three fixed directions over large areas.

Making allowance for the difference in conditions and in magnification, the pro-eutectoid cementite in these figures corresponds about as closely as could be expected to that shown in the hyper-eutectoid steel of Figs. B to M, Plate 9. Perhaps the resemblance is closest to Fig. H of Plate 9, in which the cementite is thicker than in the others, and shows clearly the strict parallelism to the cleavages of the austenite grain in which it occurs. Another point of resemblance is the irregularity with which the pro-eutectoid cementite is distributed over what was a given austenite grain.

273. The Microstructure Obscured by Pearlite Divorce.—In Figs. D, E, and F of Plate 11 we do not find the lamellar pearlite which might have been expected in these specimens because they have cooled slowly to below $Ar1$. It has evidently been broken up, either congenitally or post-genitally, by what is called "pearlite divorce," that is the disentangling of the several individual lamellæ of the pearlitic ferrite and cementite from each other's embrace. This disentangling is referred to surface tension, the effort of the retractive envelope of each mass to shrink, and thereby to bring the shape of the mass which it encloses toward the spherical, with the minimum of surface.

In much the same way the pro-eutectoid cementite is very prone to coalesce with the eutectic cementite by surface tension, perhaps aided by re-solution and re-precipitation. This pro-eutectoid cementite naturally balls up the more easily in cast iron because of the high temperature, 1,135°, at which its generation begins, and it conceals itself the more readily in cast iron because of the usually very short distances which it has to move in order to reach the adjoining masses of eutectic cementite.

Indeed it is only rarely that the pro-eutectoid cementite of the primaustenoid can be detected in cast iron.

The divorce of the pearlite may yield either *granular pearlite*, with islands of cementite in a matrix of ferrite, or else *reefs* or atolls of cementite about small islands of ferrite. If the divorced cementite succeeds in joining the eutectic cementite, its departure leaves what is essentially a ground mass of ferrite. We may now consider certain of these cases.

Many of the white cementite islands, for instance those in Fig. E of Plate 11, have probably resulted from the coalescence of divorced pearlitic cementite with the pro-eutectoid cementite, and one hesitates to say of any individual cementite lamella that it is pearlitic rather than pro-eutectoid, thickened up by the later deposition of pearlitic cementite. In parts of

Fig. F the balling up of the pearlitic cementite into granular pearlite is evident, and in parts of Fig. C of Plate 12 the ground mass of the primaustenoid areas is evidently ferrite, recognized by its etch-figuring, with spheroidal islands of cementite, and cementite lamellæ.

The rounded eutectic austenoid islands of Figs. E of Plate 11 and C of Plate 12 seem to consist essentially of etch-figured ferrite, as if all their cementite, both pro-eutectoid and pearlitic, had been attracted and ejected to the outsides of these small areas, and had there coalesced with the eutectic cementite.

274. The Microstructure Obscured by Undercooling.—In many cases the proportion of eutectic austenoid is much smaller in the outer part of the eutectic masses than in their interior, as for instance in the island marked T in Fig. F of Plate 13. Having recognized this feature of this micrograph, we can next recognize it in Fig. B of Plate 13, and then with greater difficulty in the three-rayed and the four-rayed eutectic stars G in Fig. D of Plate 12, and in the irregular eutectic stars or octopi of Fig. B of Plate 10. Then we note that, in some eutectiform masses which have less than their normal quantity of eutectic austenite, that which is present is in the central parts of those masses, the outside being composed almost solely of cementite, as in Fig. C of Plate 12, and Fig. A of Plate 11. Again we note that the rim of a cementite island[1] may differ in tint from the body of the island, as for instance is true of the white island, convex upward, at the bottom of Fig. C of Plate 12. Reefs of cementite border the great beams of ferrite in the coarse Widmanstättian pattern of Fig. B of Plate 28.

This excess of cementite about the outer parts of eutectic masses may be referred in some cases to undercooling, in others to the deposition there by surface tension of the pro-eutectoid cementite of the primaustenoid and of the pearlitic cementite, and in others to both causes. Borders that are extremely thick and regular suggest undercooling, those which are thin and irregular suggest rather the deposition of such pro-eutectoid cementite. The quantity of this latter cementite must needs be very small, whereas that of the cementite due to undercooling may be very great, as is seen on comparing the narrowness of the triangle LVh in Fig. 11 (§69), with the width of the triangle LVW.

275. An Example of Undercooling Structure.—The white rings of cementite about the honeycombed islands of primaustenoid in Fig. A of Plate 14,[2] I refer to undercooling, because of their thickness and regularity, and of the attendant conditions. As I interpret the course of events, the dark primary austenite islands solidified first, and their size was exaggerated by undercooling. On the breaking up of this undercooling the white cemen-

[1] So far as I know, Benedicks was the first to point to such a rim of cementite about the eutectic in cast iron, and to refer it to the coalescence of pro-eutectoid cementite. In his particular case the regularity of the cementite band suggests undercooling. Benedicks, *Metallurgie*, 1906, vol. 3, Fig. 12 after p. 432; reproduced by the Author, *Trans. Amer. Instit. Mining Eng.*, 1913, vol. 45, Fig. 12 on p. 379.

[2] Howe and Levy, *Proc. Cleveland Institution of Engineers*, 1913–1914, Plate 7, after p. 270.

tite bands solidified rapidly against them, impoverishing the residual molten back toward and eventually to the hypertectic carbon content. When this is reached at last, the solidification of austenite is resumed, and this, accompanying the persisting solidification of cementite, gives rise to the formation of eutectic or more accurately, of "hypertectic," which of course implies the simultaneous solidification of austenite and cementite. The eutectic may be replaced by a hypertectic. (Note E, p. 625.)

In this particular case one may hold with considerable confidence that these white rings have not resulted from post-genital divorce. First the quantity of cementite, for instance that in the part P, is too great to have resulted from the divorce of pearlite. Second, the regularity of some of these rings, for instance that about the central dark primaustenoid island, is too great to have resulted from pearlite divorce, which deposits cementite masses of irregular outline. Third, these rings cannot have resulted from eutectic divorce, because such complete divorce as they show on this hypothesis could hardly be accompanied by such perfection of the honeycombed structure in the regions abutting immediately against them. Fourth, they very certainly cannot have resulted from eutectic divorce, which at best must be extremely slow, and they almost certainly cannot have resulted from pearlite divorce, for an independent reason, that the time available was clearly far too short for the former and almost certainly far too short for the latter, in the extremely rapid, indeed almost instantaneous, solidification and cooling of this specimen, which while molten was poured in a thin sheet upon thick masses of cold steel, and has not been heated since.

Undercooling may diminish the actual quantity of eutectic to any degree. Indeed in some of the micrographs of Plate 11 the formation of eutectic seems to have been prevented wholly. Any lessening of the quantity of eutectic by undercooling results in depositing about those masses of the eutectic which actually form, the cementite which the suppressed quantity of eutectic would have contained.

276. The transformations of hyper-eutectic cast iron are the same as those of eutectic cast iron, from which it differs only in containing in addition primary cementite. This constituent of course undergoes no transformation in cooling.

The transformation is represented quantitatively in Diagram VI of Fig. 24 in the same way as in Diagram V for the eutectic cast iron.

277. (D) Summary of Solidification and Transformation.—The *solidification* of molten steel and cast iron begins with the crystallizing out of primary austenite or cementite according to whether the carbon content is less or more than 4.30 per cent., *i.e.*, to whether the alloy is hypo- or hyper-eutectic, and, in case the carbon content exceeds 1.70 per cent., ends with the solidification of the eutectic, Fig. 23, p. 130.

After this solidification the solid metal undergoes a *transformation*, which converts it into a mixture of pearlite with either ferrite or cementite according to whether it has less or more than 0.90 per cent. of carbon, *i.e.*, according to whether it is hypo- or hyper-eutectoid.

278. The solidification is selective, *i.e.*, the molten mother metal progressively brings itself toward the eutectic carbon content, 4.30 per cent., by crystallizing out primary austenite or cementite according to whether it contains initially less or more than 4.30 per cent. of carbon, *i.e.*, to whether it is hypo- or hyper-eutectic. This it does at such a rate that, under conditions of equilibrium, the carbon content of the solid and of the remaining molten are measured by the abscissæ of the solidus *AE-CFD* and the liquidus *ABD* respectively, and at the end of solidification by the abscissa of *AEBC*. Our knowledge of the hyper-eutectic part of the diagram is incomplete.

279. The transformation is selective in like manner, *i.e.*, the solid mother austenite gradually brings itself to the eutectoid carbon content of 0.90 per cent., at which carbon content it is called "hardenite," by generating ferrite or cementite according to whether it contains initially less or more than 0.90 per cent. of carbon.

This it does at such a rate that the carbon content of the residual untransformed austenite at each stage in the transformation is measured by the abscissa of the liquidoid *GOSE*, by *GOS* for steels containing less than 0.90 per cent. of carbon or hypo-eutectoid steels, and by *SE* for alloys containing more than this, *i.e.*, for hyper-eutectoid ones. Each mother grain girds about itself as a cell wall part of the ferrite or cementite which it generates; masses a part in its own octahedral cleavages; and leaves scattered within itself the remainder.

When the austenite has thus brought its own carbon content to 0.90 per cent., which it does at about 725°, it transforms as a whole into the eutectoid, pearlite, without disturbing the ferrite or cementite already generated, so that the net result of the solidification plus the transformation of the solid metal is to turn it into a mixture of pearlite with either ferrite or cementite, according to whether the mass as a whole contains less or more than 0.90 per cent. of carbon.

280. The liquidus curve[1] *ABD* is that at which solidification begins. For any given alloy it begins at the temperature at which the ordinate passing through the carbon content of alloy steel cuts the liquidus. The abscissæ of the liquidus also measure the theoretical carbon content of the residual molten metal at any given temperature.

281. The solidus curve,[1] *AEBCF*, is that at which solidification theoretically ends. For any given steel or cast iron it ends at the temperature at which the ordinate passing through the carbon content of that alloy cuts the

[1] The liquidus is sometimes spoken of as the "freezing point" and the solidus as the "melting point," but in using such language we should remember that we are using "freezing point" figuratively in the sense of the first incipiency of freezing and "melting point" as the first incipiency of melting. So, too, if we call the liquidus the "upper freezing point" and the solidus the "lower freezing point" we should recognize that these terms are correct only in a figurative sense. But in dealing with these complex matters, it is better to avoid figurative terms because they are liable to mislead through our forgetting that they are true only figuratively, or, if we remember this, then through our forgetting just wherein they are figurative and wherein accurate. Instead let us use the precise and unmistakable terms "Liquidus" and "Solidus."

solidus. The abscissæ of the solidus *AE* measure the theoretical carbon content of the particles of solid austenite which, at any given temperature, are born out of the still molten metal.

Thus for given carbon content the liquidus marks the point at which in cooling down solidification begins and in heating up freezing is complete, or in short the lower limit of complete liquidity; and the solidus marks the temperature at which in cooling down solidification is complete and in heating up melting begins, or the upper limit of complete solidity.

282. The eutectic and the eutectoid are the alloys of the lowest solidification and transformation points respectively. They are strictly parallel to the eutectics and eutectoids of other series of alloys in that,

1. The carbon content of each is fixed, 4.30 and 0.90, and independent of the carbon content of the alloy as a whole, though affected by the presence of other elements;

2. Hence the melting and freezing point of eutectic and the transformation point of the eutectoid are independent of the carbon content of the alloy, under conditions of strict equilibrium;

3. Their carbon content is not in simple atomic ratio to that of the iron; and

4. They are conglomerates of distinct minute masses of cementite mechanically mixed, in the eutectic with austenite or gamma iron saturated with carbon, and in the eutectoid with ferrite or alpha iron.

The eutectic when solidified and thereby split into its components, cementite and austenite, is called *ledeburite;* and the eutectoid when transformed and thereby split into its components, ferrite and cementite, is called *pearlite* when microscopically resoluble, and *sorbite* when it is not, the distinction thus being of degree and not of kind.

283. Eutectic cast iron and eutectoid steel are those which have the eutectic or eutectoid carbon content of 4.30 and 0.90 per cent., and hence in solidifying or transforming yield solely the eutectic or the eutectoid.

Hyper-eutectic or *-eutectoid* means containing more than the eutectic or eutectoid percentage, and hence, in freezing or transformation, yielding primary or pro-eutectoid cementite in addition to the ledeburite or pearlite as the case may be.

Hypo-eutectic or *-eutectoid* means containing less than the eutectic or eutectoid carbon content, and hence, in freezing or transformation, generating primary austenite or pro-eutectoid ferrite, as the case may be.

284. The transformations of iron and steel refer specifically to these changes from solid austenite into pearlite with pro-eutectoid ferrite or cementite in cooling down, and the reverse changes in heating up, which occur between the lines *GOSE* and *PSK* of Figs. 23 and 24.

The transformation range means this range, whether in the slightly lower position which it occupies in cooling down or the slightly higher one which it occupies in heating up. Under favorable conditions the

heating-up range tends to coincide with the cooling down range, and under ideal conditions they would coincide exactly.

285. Boundaries of the Transformation Range.—The lines *PS*, *MO*, and *GO* of the carbon-iron diagram, Fig. 28 are called *A*1, *A*2, and *A*3, respectively; *OS* is called *A*2–3; and *SK* is called *A*1–2–3. The additional letters *c*, *r*, and *e* after *A* denote rising temperature, falling temperature, and conditions of equilibrium respectively (§215).

286. Primary austenite or cementite is that which freezes out in cooling down to or toward the eutectic freezing point, in cast iron and steel respectively.

Eutectic austenite and cementite of course are those which form in the solidification of the eutectic and jointly constitute it.

Pro-eutectoid cementite is that which is born from the primary and the eutectic austenite of hyper-eutectoid steel and of cast iron as the alloy cools through Region 5, and by its birth impoverishes both those forms of austenite and so brings them to the eutectoid carbon content of 0.90 per cent.

Pro-eutectoid ferrite is that born from the primary austenite of hypo-eutectoid steel in cooling through the transformation range. By its birth it enriches the residual austenite to the eutectoid carbon content of 0.90 per cent.

Eutectoid or pearlitic ferrite and cementite are those which jointly make up the eutectoid pearlite, in the ratio of 6½ of ferrite to 1 of cementite.

CHAPTER 11

GRAPHITIZATION

287. Summary.—The graphitization of the cementite by the reaction (4)... $Fe_3C = 3Fe + Gr$, is at all times striving to occur. Its occurrence can usually be recognized micrographically by the immediate juxtaposition of the ferrite and graphite which it yields. It is much faster at high than at low temperatures. Hence it is favored by slow cooling, in that this increases the exposure to the high temperature at and just below the eutectic freezing point, 1,135°, at which the eutectic cementite forms. It is favored also by the presence of silicon and probably of aluminum, but it is opposed by the presence of sulphur and usually by that of manganese.

The graphitization of gray iron castings is chiefly that of their eutectic cementite, but that of black-heart malleable castings includes also their pro-eutectoid and their pearlitic cementite.

Cast-iron foundry practice is based primarily on controlling the degree of graphitization, so as to leave in the product the needed proportion of combined, that is ungraphitized, carbon in the form of cementite. This ungraphitized cementite is mainly pearlitic in very soft castings, but pearlitic, pro-eutectoid, and eutectic in harder ones.

In making black-heart malleable castings the graphitization is induced at about 730°, because at this relatively low temperature the mass is so rigid that the graphite forms an incoherent impalpable powder, which thus breaks up the continuity of the mass as a whole far less than the long wide flakes which graphite forms when it is generated at or just below the eutectic solidification point of about 1,135°, as in the manufacture of gray iron castings, including the interior of chilled castings.

Other means of giving the graphite a compact instead of a wide flake form are J. E. Johnson's oxygenating process, and perhaps the cold blast charcoal blast-furnace process. So, too, silicon is thought better than slow cooling as a promoter of graphitization, because slow cooling gives the graphite a chance to spread out in broad flakes.

The graphitization of primary cementite leads to the formation of kish, that is flake graphite, which often rises above the surface of the molten iron and blows away.

288. Graphitization.—Graphite is the stable form of carbon, and cementite only a metastable form always striving to change over into graphite by the reaction

$$Fe_3C = 3Fe + Gr, \qquad (4)$$

as we have already seen (§167). This graphitization complicates the solidi-

fication and transformation of cast iron, and it has to be prevented lest it complicate the transformation of hyper-eutectoid steel.

Though it is possible that graphite may, under certain conditions, form directly from molten iron or from austenite, in the majority of cases it probably forms from this decomposition of solid cementite. Hence the simpler working hypothesis, which may well be adopted, is that it forms thus in any given case. That the metastable state, cementite, should thus be the one first entered into, giving way later to the stable one, graphite, is the general law for substances which have two such states.[1]

289. Graphitization shown Micrographically.—Graphitization is often recognized micrographically by the juxtaposition of its twin products, graphite and ferrite. Because the cementite from which both spring contains $(100 - 6.67) \div 6.67 = 14$ times as much iron as carbon, and because the density of iron is about $7.88 \div 2.255 = 3.5$ times as great as that of graphite, there should be about $14 \div 3.5 = 4$ times as much ferrite as graphite by volume. Hence the area of such ferrite masses disclosed in a microsection should be about 4 times that of the graphite[2] against which they abut (§123).

This juxtaposition of the black graphite and the white ferrite is shown in Figs. E of Plate 4, C and E of Plate 5, F of Plate 6, D and G of Plate 7, and F of Plate 12. In Figs. D and G of Plate 7 each little black graphite crater[2] is surrounded by a white ring of ferrite. In the other figures cited the sinuous black lines or bands represent graphite, and the conformable broad white bands which border one or both sides of these graphite masses are of its twin product, ferrite.

290. Conditions Favoring the Graphitization of Cementite.—The tendency of the cementite to graphitize increases

1. With the temperature;
2. And hence necessarily with the time available at high temperatures, that is with the slowness of cooling; and,
3. With the content of silicon, and probably of aluminum.
4. It is greatly strengthened by the presence of graphite nuclei. On the other hand it is opposed
5. By the presence of sulphur, and
6. By that of maganese under most conditions
7. There is much to suggest that increasing carbon content, besides its certain indirect effect of raising the temperature at which cementite comes into existence, and thus acting through influence (1), also directly increases the graphitizing tendency.

The direct effect of manganese is to oppose graphitization, because the exothermic manganese carbide, Mn_3C, unites isomorphously with the

[1] Ostwald, "Lehrbuch der Allgemeinen Chemie," Engelman, Leipsic, 1902, vol. II, 2, p. 444.

[2] Very often these black areas and lines which represent graphite are actually cavities left by the accidental removal of the friable graphite in the process of polishing the specimen for microscopic examination. In some cases at least the graphite areas contain metallic skeletons.

iron carbide to form a manganiferous cementite $(Fe,Mn)_3C$, which is much more stable than a pure iron cementite, and thus resists graphitization much better. But in addition manganese has some other and opposite effect which may outweigh this under certain conditions.

In one sense the influence of the silicon, manganese, and sulphur content is predominant. That is to say, if silicon is wholly or nearly wholly absent, or if the content of either manganese or sulphur is so great that its effect far outweighs that of the silicon present, then graphitization may be extremely slow even at temperatures as high as 1,050°.[1]

But the influence of temperature is very marked under other conditions. Thus even in the absence of silicon, graphitization is not only very rapid at temperatures as high as 1,135° but accelerates with any further rise of temperature (§293). And in the presence of even a moderate silicon content both the temperature and the time available for graphitization have great influence.

The influence of temperature and of carbon content jointly is shown by the increase of the tendency to graphitize with increasing carbon content. Thus hypo-eutectoid steel with less than 0.90 per cent. of carbon, generates cementite only at temperatures as low as about 700°, and hence shows very little tendency to graphitize. In hyper-eutectoid steel the tendency to graphitize increases with its carbon content, with which the temperature at which the cementite is generated rises from about 700° with 0.90 per cent. of carbon to 1,135° with 1.70 per cent. In cast iron, in which there is abundant deposition of cementite at 1,135°, the tendency to graphitize is proportionally stronger.

Here, too, the influence of silicon is shown by the rapidity with which graphite forms even in relatively low-carbon steel if the silicon content is great. Thus Charpy and Cornu-Thenard[2] find that, though when the carbon content is between 0.14 and 0.40 per cent., no graphite forms at any temperature in the presence of only 2.20 per cent. of silicon, illustrating the influence of the small carbon content, yet if the silicon rises to about 4 per cent., graphitization is complete in 3 hours at 800°, and nearly complete in 1 hour, even if only 0.14 to 0.15 per cent. of carbon is present. Their inference, that graphitization is so greatly stimulated by the presence of much silicon that it may occur fast enough to cause a detectable retardation of the cooling, is of great interest. (See Lissner, *Ferrum*, 1912, vol. 10, p. 52 and Plate II.)

The influence of time is seen in the manufacture of chilled castings, of which the rapidly cooling face has very much less graphite than the slower cooling interior.

[1] Thus Tiemann found that exposures even of 1 hour at 1,116° and of 3 hours at 1,012°, caused no appreciable graphitization in a white cast iron containing initially no silicon, 4.271 per cent. of combined carbon, and 0.255 per cent. of graphite, or together 4.526 per cent. (*The Metallographist*, 1901, vol. 4, p. 319). Again the various specimens of white cast iron shown in Figs. A to F of Plate 11 and C of Plate 12 of this book, which were held for an hour or more at about 1,050°, gave no indication of graphitization. Their silicon content is about 0.10 per cent.

[2] *Journ. Iron and Steel Institute*, 1915, No. I, Vol. 91, p. 276.

The stimulating effect of the presence of graphite on graphitization, this self-acceleration of graphitization, may be traced in connection with exposures to constant and to falling temperature, as in the manufacture of malleable castings. But in connection with exposures to rising temperature it is less evident, because here the increasing solubility of carbon in the austenite leads to the dissolving of part of the graphite present, which precipitates as cementite when the temperature falls again. On this account exposures to nominally constant temperature may transform graphite into cementite, because these temperatures are not truly constant but oscillating, and at every slight rise some of the graphite dissolves in the austenite, to re-precipitate as cementite during the ensuing slight fall.

That the presence of graphite nuclei should stimulate graphitization is only a single application of the general law that the presence of nuclei of the stable form hastens the change to this form from the metastable form, as nuclei of sodium sulphate dropped into a supersaturated solution induce extremely rapid crystallization.

291. Cast-iron foundry practice is based primarily on controlling the degree of this graphitization of cementite. The conditions under which the cast iron is made in the iron blast furnace give it an initial content of carbon, and hence of the brittle glass-hard cementite, far greater than can be tolerated for most industrial purposes. Reaction (4), p. 215, causes a radical change in the properties of the mass, substituting for cementite two extremely soft substances, ferrite and graphite, the former fairly strong, the latter extremely weak and incoherent, indeed practically a foreign body which only weakens the mass by breaking up its continuity.

But while the initial hardness due to the presence of the cementite can be almost completely effaced as in the manufacture of black-heart malleable castings by thus substituting for cementite these two extremely soft substances, ferrite and graphite, the initial brittleness can be removed only in part, because the loss of continuity caused by the presence of graphite necessarily prevents any great degree of malleableness. Even the removal of this graphite by oxidation does not help matters greatly, because that removal does not close up the spaces which the graphite has occupied, and because these spaces are almost as effective as the graphite itself in destroying malleableness.

This reaction is controlled in the foundry so as to give a ratio of ungraphitized cementite to graphite + ferrite appropriate to the conditions of service for which the castings are to be used. In making chilled castings and others needing great hardness, graphitization is purposely hindered; while, as explained in §§39 and 128, for making "malleable" castings and others which need softness, in order that they may be machined cheaply, and such slight malleableness as can be given, this reaction is stimulated, and may indeed be pushed to completion, so that the whole of the cementite is graphitized.

This control of the degree of graphitization is brought about chiefly by

regulating (2) and (3), the length of time available for graphitization and the percentage of silicon. This latter must of course be governed by (5) and (6), the quantity of manganese and of sulphur present which the silicon has to counteract. Only very rarely, as in making certain very hard cast-iron projectiles, is the sulphur content purposely increased in order to restrain graphitization.

292. The length of time available for graphitization, (2), is in turn controlled chiefly (A) by the thickness of the casting itself, a thick casting spontaneously cooling more slowly than a thin one; (B) by the nature of the mould, as in the manufacture of chilled car wheels and other chilled cast iron objects, the part of the mould against which the tread of the wheel solidifies being itself of cold iron which conducts the heat away quickly, whereas the remainder of the mould is made of sand and clay which conduct the heat relatively slowly; and (C) by artificial long heating, as in the manufacture of malleable castings.

293. That the Rate of Graphitization Increases with the Temperature Ought to be Familiar to All.—Those who practise the malleable iron industry report that graphitizing goes on very much faster at 870° than at 825°. Graphitizing occupies fewer minutes at the eutectic freezing point, about 1,135°, than days at the industrial annealing temperature of the malleablizing process. Ruff[1] assumed that the time occupied by graphitizing half the cementite present was decreased by half for every 10° rise of temperature just below 1,135°, being 16″ at 1,105°, 8″ at 1,115°, and 2″ at 1,135°. Even if these numbers are only very roughly true, they imply, for the much higher temperatures at which the precipitation of pro-eutectic cementite begins in strongly hyper-eutectic cast irons, a practically instantaneous graphitization of the cementite, a graphitization so rapid that no appreciable quantity of ungraphitized cementite should be present at any instant in the molten metal during the early part of the kish formation.

Further evidence that the rate of graphitization increases with the temperature is at hand in the slighter graphitization (1) of steel than of cast iron, and (2) in hypo- than in hyper-eutectic cast iron, as shown by the kish phenomena and like evidence.

(1) In the solidification of cast iron, which must needs always contain eutectic, the generation of solid cementite begins with the eutectic solidification, *i.e.*, at about 1,135°, whereas in the solidification of steel containing less than 1.70 per cent. of carbon no cementite forms till the temperature falls to the line *SE*. Even after all possible allowance is made for the stimulation of graphitization by the presence of silicon in cast iron, the graphitizing tendency of steel still falls so far behind that of cast iron that much of the difference refers itself to the lower temperature at which the cementite of steel comes into being. So, too, the common observation that even among hyper-eutectoid steels, the graphitizing tendency increases with the carbon content, explains itself as reflecting the rapid rise of the line *SE*. That is

[1] *Metallurgie*, 1911, vol 8, p. 507.

to say, whereas in a steel but slightly hyper-eutectoid, with say 0.95 per cent. of carbon, no cementite forms till the temperature sinks nearly to 725°, and none remains if the temperature rises materially above 725°, in one which is strongly hyper-eutectic, containing say 1.70 per cent. of carbon, pro-eutectoid cementite begins separating as soon as the temperature starts to pass below 1,135° during solidification, and remains always present at all temperatures below this, unless indeed it has already graphitized.

294. What Cementite is it that Graphitizes?—Because the rapidity of graphitization decreases extremely rapidly as the temperature falls (§293) it is only at the temperature of solidification, about 1,135°, and immediately below it that graphitization has time to go on to any important degree in the relatively rapid cooling of most gray cast-iron objects. Hence it must be chiefly the eutectic cementite that graphitizes, because at these temperatures there is little other cementite which has crystallized out from solution and is therefore in condition to graphitize readily. There is no primary cementite because foundry castings are not rich enough in carbon to be hyper-eutectic, and hence they form none. There is but little pro-eutectoid cementite at these temperatures, first because the total quantity of this form of cementite which can form is very small, as is seen on noting how narrow the triangles *Rwt'* and *LVh* in Fig. 11 are; and second the generation of even this little is spread out over the whole range from 1,135° to 725°, and in all but the very upper part of this range graphitization is slow, so that most of this little is generated at temperatures too low to permit it to graphitize to any important degree in the relatively rapid cooling of most gray iron castings. The low temperature, 725°, at which the pearlitic cementite is generated is an effective bar to its graphitization in such a cooling.

Internal evidence of this graphitization of the eutectoid cementite is given in Fig. C of Plate 5. The outline of the large white mass marked T, T, T, is evidently that of a eutectic, like that of T in Fig. E of Plate 4, and that of T in Fig. F of Plate 13. Here the eutectic has been converted into two masses, one of graphite and one of ferrite, which remain conformable to the initial outline. It is a case of paramorphism, or change by re-arrangement in the proximate constitution of the body without change in its ultimate composition or material change in its outward form.

295. An Example of Graphitization at Relatively Low Temperatures.—But though graphitization becomes slower and slower as the temperature falls, it still goes on, apparently even at temperatures as low as the bottom of the transformation range, *Ar*1 or 725°. Thus in the very slow cooling of the malleable casting process even the pearlitic cementite graphitizes.

A comparison of Figs. D and E of Plate 5 throws light on this low-temperature graphitization. These show two like pieces of gray air-furnace cast iron, both re-heated to 900° for 1 hour, and then cooled, E very slowly in the furnace, D relatively fast in the air. The white masses of ferrite bordering the graphite flakes of E are replaced in D by a gray ground mass, through which run many thin curved lighter gray streaks, such as X. Sodium

picrate darkens these, proving that they are of cementite. This enables us to interpret the contrast between E and D in the way which I will now describe.

The solvent power of the austenite of the casting is greater at 900° than at 725°. Hence when the casting was heated to 900°, part of its austenite began dissolving carbon from any neighboring masses of either graphite or cementite. This re-solution of graphite on reheating has been proved abundantly, and is indeed the reason why malleable castings are so extremely difficult to forge when hot. Heating them converts them into a mixture of graphite with steel. This steel increases rapidly in carbon content as the re-solution of the graphite continues, and soon becomes hyper-eutectoid, and on the slightest cooling again precipitates pro-eutectoid cementite.[1]

In the subsequent air cooling of Fig. D the decrease of solubility of carbon from 900° to Ar1, 725°, caused a re-precipitation of the excess of carbon which had been dissolved in heating from 725° to 900°. This re-precipitation takes place not as graphite but as cementite, usually if not always, in conformity with Ostwald's law that the metastable state is that first entered. In this relatively rapid cooling there was no time for the graphitization of these cementite masses, much less was there time for the graphitization of the pearlitic cementite. Hence the occurrence in Fig. D of these pro-eutectoid cementite masses in a pearlitic or sorbitic ground mass.

In the case of Fig. E a like dissolving of carbon took place in the heating to 900,° and a like precipitation of carbon took place on again cooling down; but now the very slow cooling affords time for the graphitization of the cementite thus precipitated, so that not only are these pro-eutectoid cementite masses lacking, but the pearlitic gray ground mass of Fig. D is here replaced for the most part by white ferrite areas with dark graphite flakes. This means that during this slow cooling there has been time, not only for the pearlite to graphitize and thus turn into a mixture of ferrite plus graphite, but that the graphite thus precipitated has had time to coalesce into its characteristic curved flakes.

This is true of the upper and lower parts of Fig. E, but in the central part this process has not gone so far, and many gray pearlitic areas persist between the ferrite areas. This grayish central part of the micrograph thus represents a stage of graphitization intermediate between that of Fig. D in which there has been little graphitization in this last cooling, and that of the whiter parts of E in which the graphitization is almost complete.

The curved shape of the pro-eutectoid cementite masses contrasts with their usual straightness, as in Figs. C to L of Plate 9. It may mean that these masses reproduce the shape of the prior graphite masses, which dissolved in the heating to 900°. As when dissolved they began diffusing away, the thin area where they were would be the slowest to lose its carbon

[1] See Benedicks, *Metallurgie*, 1908, vol. 5, p. 41; and the Author, *Trans. Amer. Inst. Mining Eng.*, 1912, vol. 44, p. 470, Table III and Fig. 7.

by diffusion, and thus would remain the richest in carbon, and hence in cooling would earliest become super-saturated with carbon, and hence would be the area in which precipitation of pro-eutectoid cementite would begin. To these nuclei, standing though they do in this curved line, the additional cementite, precipitated during the further cooling, would migrate after the manner of precipitating salts, thus reproducing approximately the shape of the initial graphite flakes.

The graphite is more prominent in Fig. E than in Fig. D partly because there is more of it, thanks to the graphitization in the slow cooling of E, but also in important part because the black graphite contrasts more strongly with the white ferrite than with the gray pearlite of Fig. D. The strong contrast in Fig. E over-accents the graphite, the feeble contrast in Fig. D tends to conceal it.

The relatively rapid graphitization of Fig. D should be referred to the presence of much graphite when this second cooling started.

The reason why the graphite precipitated in this cooling does not form incoherent compact masses as in the annealing of the malleable casting process probably is that the presence of the graphite flakes formed in the initial solidification determines the precipitation of the graphite against them, and its identifying itself with them.

296. Mechanism of the Migration of Carbon from a Cementite Mass to a Graphite Mass by Solid Sublimation.—From the fact that graphite is the more stable and cementite the less stable or metastable form, it follows that the solubility of carbon in austenite, at a given temperature, must be less in the immediate neighborhood of a graphite mass than in that of a cementite mass, in agreement with the general law that the solubility depends on the nature of the "Bodenkörper" or substance with which the solution is in contact as will be explained later, and is less in contact with the stable than in contact with the metastable form. This gives us the key to the migration of the carbon from the cementite masses to those of graphite.

Consider a mass A of cementite in a ground mass of austenite at 1,000° as sketched in Fig. 32. Let us suppose that the cementite has followed its natural course and dissolved in part in the surrounding layers of austenite, 1 to 5, till it has saturated those layers with Y per cent. of carbon, that which suffices to saturate austenite when in contact with cementite. Now let us suppose that a mass of graphite B is introduced. Any carbon dissolved in layer 5 of the austenite in the immediate neighborhood of B, in excess of the percentage, Z, which suffices to saturate austenite when in contact with graphite, tends to precipitate out. Being in immediate contact with a nucleus of graphite it naturally precipitates as graphite. When this reduction of the graphite carbon content of layer 5 from Y to Z per cent. occurs, the carbon in layer 4 immediately begins to diffuse into 5, quite as the salt from a more concentrated part of a liquor tends immediately to diffuse out into a neighboring poorer part of that same liquor. Concentration always tends to become uniform, quite as water tends to become level. But the migra-

tion of carbon from 4 to 5, immediately makes 4 poorer in carbon than 3, and this inequality starts carbon migrating from 3 into 4, and by like process starts migration from each layer to that on its right, till at last the carbon content of layer 1 is somewhat reduced. But this reduction brings its carbon content below Y, that which it is able to dissolve from A, with the result that more carbon immediately passes from A into 1.

But this raising of layer 1 to or toward Y per cent., while layer 5 is having its carbon content drained down to or toward Z per cent., re-establishes the carbon gradient which diffusion was trying to flatten out, and thus stimulates diffusion afresh, which goes on in the vain attempt to equalize the carbon content of the austenite from 1 to 5 in spite of the fact that B drains 5 down to Z per cent. while A crams 1 up to Y per cent. Manifestly there is no end to this before the whole of A has dissolved, and all of its carbon, and also all the carbon initially present in the austenite in excess of Z per cent., has migrated across to B, and precipitated on B as graphite.

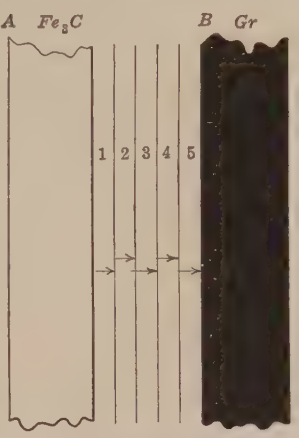


FIG. 32.—Mechanism of the migration of carbon from a cementite mass to a graphite mass by sublimation.

The process is strictly analogous to sublimation, the transfer of substance from one solid to another, save only that the medium through which this transfer occurs is solid yellow-hot austenite in our present case, instead of a gas as in our familiar cases of sublimation.

This conception helps us to understand why the presence of graphite nuclei hastens graphitization. Every particle of graphite acts as a center, sucking to itself as it were the carbon of the surrounding austenite, to replace which the neighboring particles of cementite start dissolving in that austenite, and so much as thus dissolves migrates across the austenite to the graphite nuclei. As the surface of these grows by accretion their sucking action increases proportionally.

297. Control of Graphitization in making Black Heart Malleable Castings.—When graphite is generated at the high temperature of solidification, about 1,135°, and at temperatures immediately below, the mobility of its environment is so great that it readily assembles into long thin flakes which break up the continuity of the mass very seriously, as in even the excellent gray cast iron castings shown in Figs. G of Plate 6 and F of Plate 7. Of these the latter shows a good gun-metal casting. (See §302.) If the graphite is generated at a much lower temperature, say at 730°, it finds itself in so rigid an environment that it forms only an impalpable powder filling the interstices between a skeleton of ferrite or of pearlite, as is shown in Figs. A and B of Plate 4, and C of Plate 7, and thus breaks up the continuity much less harmfully. The purpose of the black heart malleable process is therefore to avoid generating graphite during and immediately after solidification, and instead to induce it at about 730°.

Hence in this process the cast iron when poured has such a ratio of manganese + sulphur on one hand to silicon on the other as shall restrain the graphitizing tendency so fully that no important quantity of graphite forms during the relatively rapid solidification of these usually thin and hence fast-cooling objects; and yet not so fully but that the whole of the cementite can be graphitized later by reheating these castings to about 730°, and holding them at this temperature for about 60 hours (§ 39).

298. Control of Graphitization in making Gray Iron Castings.—These usually have no thermal treatment after their initial casting and cooling down in the mould, and hence no later opportunity for graphitization. Hence their ratio of silicon to manganese plus sulphur must be regulated so as to induce at that time the desired degree of graphitization, thickness of casting and consequent rate of cooling being taken into account.

But in effect the rate of graphitization falls so extremely rapidly with the temperature, that whatever graphitization there is must needs occur during and immediately after solidification, when the solid metal is so mobile that the nascent graphite assembles in the relatively large flakes which, forming a nearly continuous skeleton, cover the whole fracture of a good gray casting.

Were this skeleton absolutely continuous, then the strength of the casting would be only that of graphite. In fact the casting owes its strength mainly to the discontinuities in this graphite skeleton.

299. J. E. Johnson's Process of Spheroidizing Graphite in Cast Iron.[1] **(See §124).**—He spheroidizes the graphite of cast iron by bessemerizing the iron at a relatively low temperature, and thus removing its silicon with but little removal of carbon but with important absorption of oxygen. The graphite of the resultant product assembles in large part in relatively harmless spheroids, instead of spreading out as usual into broad continuity-destroying flakes. He then re-melts his iron, adding enough silicon to bring about the degree of graphitization desired in the castings themselves. The spheroidizing tendency persists through this re-melting, so that the graphite in the resulting castings also is chiefly in compact masses, such as are shown in Fig. F of Plate 4.

300. Theory of this Process.—The process is a result of his search for the reason why some cast iron which he made in charcoal blast furnaces was so surprisingly strong. The graphite of these irons was in relatively compact masses, and to this compactness he referred their strength. But why this compactness? His observations connected it with cool-running in the blast furnace. But why should cool-running cause compactness? Oxygenation of the iron suggested itself as the immediate cause. Seeking a reason for this cause, he conceives that the presence of oxygen in the cast iron lowers the temperature of graphitization relatively to that of solidification, with the result that the graphite, when it comes into existence, finds itself

[1] J. E. Johnson, Jr., *Trans. Amer. Inst. Mining Eng.*, 1912, vol. 44, pp. 314; and 1914, vol. 50, p. 344; *Iron Age*, 1912, vol. 89, p. 1206, and vol. 90, p. 1375.

imprisoned within a relatively rigid mass, and hence has but little power to spread out into broad flakes, but instead forms relatively compact spheroids.

The greater compatibility, if I may say so, of oxygen and carbon at lower than at higher temperatures is indeed a most familiar phenomenon of steel metallurgy. For instance in the Bessemer process an elevation of the temperature leads to the oxidation of carbon in preference to silicon, and a depression of the temperature to the opposite effect. In the acid open hearth and electric processes, too, silicon may be returned from slag to metal[1] by raising the temperature, a phenomenon interpreted as meaning that the rise of temperature favors the oxidation of carbon even at the expense of silicon, by the familiar reaction (15)....$SiO_2 + 2C = Si + 2CO$, a reaction which occurs readily in the presence of molten metallic iron.

Granting this compatibility, the reason why the iron from a cool-running furnace should contain oxygen is near at hand. As the molten iron, after trickling down through the rapid inrush of relatively oxidizing gases at the tuyére level, first assembles in the hearth, it may indeed be relatively rich in oxygen, because till now it has encountered no very strong deoxidizing action. It is the projection of the bottom of the column of incandescent fuel into the bath of molten iron in the hearth of the furnace, *L* of Fig. 3, that gives the really strong deoxidizing action of the blast-furnace process. But if the hearth is cool, this deoxidizing is correspondingly weak. Hence the oxygen-content of irons made in cool-running.

At these relatively low temperatures appreciable quantities of oxygen and of carbon can coexist. Johnson reports cases in which as much as 0.065 per cent. of oxygen coexists in cast iron with 4.10 per cent. of carbon, 0.70 per cent. of silicon, and 0.60 per cent. of manganese,[2] E of his Table I.

301. Supporting Evidence from the Charcoal Blast-furnace Process.— In support of his contention he points to the low hearth-temperature of charcoal blast furnaces, especially when they are run with cold blast, compared with that of coke furnaces, and to the greater strength of the iron made in these cool hearths. The fusible slags of the charcoal furnace permit running at a low temperature, and thus avoiding the high temperature which would deoxidize the product thoroughly. In coke furnaces, on the other hand, the slag must be very calcareous in order to absorb the sulphur brought in by the coke, and because calcareous it must be relatively infusible. Because it is relatively infusible the hearth temperature must be very high in order to give the needed fluidity, and this high temperature deoxidizes the cast iron thoroughly, and thus leads in turn to the spreading out of its graphite in broad flakes. In order that the strength of such irons should support his hypothesis it must be shown by abundant evidence to be accompanied by compactness of the graphite masses, referable to these

[1] For a late confirmation of this reduction in the acid electric furnace see A. Müller, *Stahl und Eisen*, 1914, vol. 34, p. 89, and also *Iron Age*, 1914, vol. 93, p. 670.

[2] *Trans. Amer. Inst. Mining Eng.*, 1914, vol. 50, p. 354.

hearth conditions. Examining 11 specimens of American cold blast charcoal pig iron, apparently from three makers, I find only two in which the graphite is appreciably more compact than in coke irons, and these two are graded Nos. 4 and 5. Fig. A of Plate 5 shows about the usual structure of the graphite in the nine remaining specimens. Six such irons of Swedish source indeed had very compact graphite, as shown in Fig. AA of Plate 5. Unfortunately there is reason to suspect that in all these cases this compactness of graphite was accompanied by an unusually small graphite content, which would suffice to explain the compactness, and thus prevent our referring it to special hearth conditions. These observations, while insufficient to disprove Johnson's hypothesis, yet incline us to exact full evidence in its support.

To many the weakest point in this theory will seem the persistent coexistence of such considerable quantities of oxygen on one hand and of readily oxidizable elements on the other, carbon, silicon, and manganese, in molten cast iron. But this has been established by so many analyses that it is not open to question. Indeed the coexistence of very appreciable quantities of oxygen and these oxidizable elements has long been known, for instance, 0.782 per cent. of oxygen with 0.083 per cent. of manganese reported by Müller, and 0.187 per cent. of oxygen with 0.123 per cent. of carbon reported by Ledebur.[1] Indeed the value of oxygen as a constituent of cast-iron castings is so firmly accepted that some of the most intelligent founders determine the oxygen content regularly, and apparently are guided by it in selecting their iron. The oxygen content which Johnson's process is now able to give is inversely as the silicon content, decreasing from about 0.06 per cent. to a very small quantity as the silicon content rises from 1 to 3 per cent.

The process has already proved its technical merits so far that its product has reached the thousands of tons.[2]

302. The influence of the rate of cooling on the quantity and size of the graphite particles is shown by Figs. F and G of Plate 6, and Figs. A to F of Plate 13.

Fig. G of Plate 6 shows the corner and Fig. F of Plate 6 the interior of a pyramid of good gray foundry cast iron. Note that in the faster cooling corner the graphite masses are fewer and smaller than in the slower cooling interior. Here the sole difference in conditions is in the rate of cooling. Indeed similar variations in the coarseness and abundance of the graphite can be brought about even among castings only $\frac{1}{2}$ in. in diameter, poured in sand moulds from the same molten mass and within a short time of each other, by varying the rate at which they cool. This can be done by such expedients as varying the dampness of the moulding sand, varying its compactness by varying the firmness with which it is rammed, and varying the temperature of the metal by holding it longer in the casting ladle.

Figs. A to F of Plate 13 show the microstructure of various parts of one

[1] The Author, "The Metallurgy of Steel," 1890, Table 43, p. 94.

[2] Private communications, J. E. Johnson, Jr., Nov. 16 and 18, 1915.

of Mr. A. W. Whitney's unusually good chilled cast-iron railroad car wheels, of which he has kindly sent me fragments.

303. Conditions in Making Chilled Cast-iron Car Wheels.—In making these car wheels the part of the mould against which the tread, that is the outer face of the rim, or wearing surface, solidifies, is made of cast iron, and the rest of the mould is made of moulding sand. The purpose of this is first to solidify and cool that wearing surface rapidly, so as to give little time for graphitization, and thus to retain most of the carbon in the condition of cementite, and thus to give this surface the hardness to resist the abrasion of the rails and especially of the brake shoe. Second, to cool the remainder of the wheel, including both the interior parts of the rim, the hub, and the "plate," which is the member connecting the rim with the hub, slowly enough to give time for the graphitization of the cementite. Thus in the plate the ductile ferrite, mixed with the unbrittle graphite, is substituted for the hard brittle cementite which, if it were allowed to persist ungraphitized, would make this part of the wheel far too brittle.

Then the wheels are removed from their moulds while still red hot, piled in a pit, and allowed to cool very slowly, thus losing the casting stresses, and having their austenite changed into pearlite. Thus is made a wheel with a wearing surface which is very hard because of its large content of cementite, integrally united with a body which has surprisingly good power of enduring shocks.

304. Difference in Microstructure in Different Parts of a Chilled Cast-iron Car Wheel.—Figs. A and B of Plate 13 show the wearing surface, magnified 47 and 500 diameters, C and D the mottled region about 1 in. back from the face surface, and E and F gray "plate." Note the great contrast between A and E. The former consists of the dark well-oriented pine-tree masses of dark primaustenoid, resolved in Fig. B into pearlite in a matrix of white eutectic. In Fig. E, on the contrary, only relatively few crescent, spandrel, or concave white eutectiform islands, T, persist, most of the eutectic cementite having been resolved by reaction (4) into the sinuous black graphite flakes, imbedded in a sea of pearlite resulting from the transformation of the primary austenite and the eutectic austenite in cooling. The pro-eutectic cementite generated along with this pearlite in the cooling of this austenite has either graphitized or coalesced with the small residue of ungraphitized eutectic cementite.

The relation of the small quantities of eutectic which have escaped graphitization, to the accompanying primaustenoid is shown in Fig. F of Plate 13. This primaustenoid, or mixture of pearlite and pro-eutectoid cementite into which the primary austenite has transformed in cooling (§80), has in Fig. F a bold convex promontory marked S, pointing as it were northwest, and jutting boldly into the concavity in the large light spotted island of eutectic marked T. Here the pearlite is readily identified by its characteristic lamellar structure, as well as by the shape of this promontory inherited from the primary austenite; and the eutectic is equally well

identified both by its honeycombed structure, and by its concavity to this pearlite mass.

The graphitization in the mottled part of this same casting, Fig. C, Plate 13, is interesting. The dark circular area in the center of the micrograph is one of the black spots usually seen by the naked eye in these mottled regions. On ocular inspection this spot might be supposed to represent a spherical body of graphite, but closer examination, especially under the larger magnification of Fig. D, leads us to a different interpretation. First we should bear in mind that the rounded dark areas in malleable castings form far too great a proportion of the whole to represent graphite, and hence presumably represent spheroids spanned by thin ferrite skeletons, and filled in with fine particles of graphite. Next we should note the relatively narrow pearlite areas between the graphite masses in Fig. D.[1] With these things in mind we interpret the lighter parts of the dark circular area in Fig. C as representing this pearlite, which naturally colors darker than the cementite and eutectic masses in the remainder of the field, though lighter than the graphite. We interpret the darker parts of this area as representing in part graphite, and in part spaces left by the breaking down of these thin pearlite partitions during the polishing and etching. They would be likely thus to break down, both because they are relatively thin, and because they are much softer than the white area outside, which is hard because of its great content of eutectic cementite.

It is interesting to note in Fig. D that in this mottled cast iron too, a part of the graphite forms sinuous flakes, in a general way like those of the gray iron shown in Fig. E.

305. Difference between the Dark Spots in Mottled Castings and the Graphitic Areas in Malleable Castings.—These dark areas in mottled iron differ from those in malleable castings, discussed in the middle of §123, Chapter 6, not only in having formed at a much higher temperature, during or shortly after solidification, instead of at the relatively low temperature, about 730°C., of the annealing process for malleable castings, but also in representing much less thorough graphitization. Thus the pearlitic cementite shows no sign of having graphitized in Fig. D of Plate 13, whereas in properly annealed malleable castings the pearlite is almost completely graphitized, so that the mass consists almost solely of ferrite and graphite.

We have already seen that in the annealing of malleable castings, the eutectic cementite is reported to graphitize before the pearlitic,[2] so that at a stage in that process before that shown in Fig. C, of Plate 7, pearlite masses corresponding to those of Fig. D of Plate 13 should be found.

306. Advantage of Silicon over Slow Cooling as a Promoter of Graphitization.—If a given degree of graphitization is sought, it is clearly less harmful

[1] Dr. J. E. Stead informs me that he finds this same structure in mottled Swedish cast iron. "In the whiter grades of metal the graphite plates are always in juxtaposition with pearlite." Private communication, July 27, 1914.

[2] E. Touceda, private communication, July 3, 1914.

if brought about by an abundance of silicon coupled with less slow cooling than if caused by slower cooling with less silicon. For the slower cooling, by lengthening the time of high mobility, increases the opportunity of the graphite to assemble in broad flakes, which are more destructive of continuity than the narrower and shorter flakes to which a less slow cooling leads, as pointed out by Sauveur.[1]

How great the difference in the size of the graphite flakes may be in commercial castings is seen on comparing Figs. E and F of Plate 12, which have the same magnification. Fig. E, with its shorter graphite flakes, is of good radiator cast iron; Fig. F, with its coarser flakes, is of bad coke car-wheel cast iron, described by J. E. Johnson, Jr.[2] The thin walls of the radiators necessarily cool rather rapidly, so that graphitization sufficient to give not only the softness needed for machining, but also shock-resistance, has to be brought about by a rather high silicon content, about 2.25 per cent. Such a length of graphite flakes as is shown in Fig. F would be very harmful here, in causing porosity, and hence leakage.

This contrast between Figs. E and F remains striking even after we allow for the optical illusion which exaggerates the width but not the length of the graphite flakes of Fig. F. In polishing the section the graphitic area, because softer, naturally grooves out, and the edges of the ferrite adjoining this area are thus made to slope down toward the groove. Because they slope they do not reflect their share of light into the camera and hence photograph dark.

307. The kish and primary cementite phenomena, item (2) of this present itemizing, in like manner relate themselves to the hastening of graphitization with rise of temperature (§293). In fact primary cementite is rarely, if ever, present in commercial cast iron, unless its graphitization has been prevented specially, as by the presence of much manganese or by quenching the molten metal in water. Moreover, so acute an observer as Mr. J. E. Johnson finds that most kish-bearing cast irons, *i.e.*, those which give off kish before solidifying, are unusually fluid.

The absence of primary cementite from commercial cast iron free from manganese explains itself as meaning that, at temperatures at which such cementite can form, necessarily above the eutectic freezing-point, 1,135°, this cementite always graphitizes completely, and hence is never found as such.[3]

This absence of primary cementite implies that the combined carbon content cannot exceed the eutectic content, 4.30 per cent. in pure iron carbon alloys. For the combined carbon is necessarily present as cementite, and primary cementite is the only form of cementite which can be present in excess of the eutectic content.

308. Kish appears to be graphite resulting from the instantaneous graphitization of primary cementite.

[1] *Trans. American Instit. Mining Eng.*, 1913, vol. 45, p. 361.

[2] *Idem*, 1913, vol. 44, p. 314.

[3] See the Author, "Does Commercial Hyper-eutectic Cast Iron Free from Manganese Exist?" *Journ. Iron and Steel Inst.*, 1912, No. II, vol. 86, p. 291. In this paper I called for instances of the existence of such iron. Thus far none have been cited.

The thin flakes of graphite which thus form are so much lighter than the molten metal that they rise quickly to the surface, and escape in large part into the air. This flaky graphite floats about the casting house, and accumulates on what might be called the "high lights," for instance on one's shoulders and on all nearly level surfaces. The exact mechanism of its effecting its escape from metal is not certain. We may suppose that, rising swiftly and on edge, it has such momentum that it is able to break through the surface tension and to bob up higher than it would in equilibrium, as a pole forcibly held below the water rises high out of the water, when released, to sink again toward an equilibrium depth. Thus kinetically protruding from the iron, the light flaky graphite may offer enough surface to the strong convection currents which play over the surface of the metal to enable them to sweep it up out of the iron and bear it away. Part of this kish is found coating the solidified pigs and other masses of iron.

If, as I infer, kish is thus a product of the graphitization of primary cementite, then strongly kish-bearing means strongly hyper-eutectic. But to become strongly hyper-eutectic the iron must needs be heated within the blast furnace to a temperature far above the eutectic freezing point, because of the steepness of the hyper-eutectic liquidus. Hence a strongly kish-bearing iron means an unusually hot iron, which because unusually hot would naturally be unusually fluid. This explains Mr. Johnson's observation just cited, that kish-bearing irons are more fluid than others.

Evidently the pro-eutectic solidification of hyper-eutectic irons is represented by this escape of kish. The fluidity during this period of solidification contrasts strongly with the mushiness during the corresponding period for steel and hypo-eutectic cast iron. This contrast is quite parallel to that in our nitrate-water series, §161.

In each case the pro-eutectic solidification of the hypo-eutectic substance is that of a bulky coherent mass, which accumulates at the surface and thrusts itself conspicuously on our attention, primary ice from our aqueous solution, primary austenite from our cast iron. It has all the characteristics of the solidifications with which we are familiar, such as those of ice, glass, most metals, and fused salts, and hence it recalls those solidifications and identifies itself as of their class.

But the corresponding pro-eutectic solidification of the hyper-eutectic nitrate and the hyper-eutectic cast iron has none of these characteristics. It is the solidification of a small, compact, incoherent substance which quickly removes itself, the nitrate or other salt which crystallizes out of a hyper-eutectic aqueous solution falling to the bottom and not interfering with the fluidity of the remaining mother-liquor, the cementite of the hyper-eutectic cast iron graphitizing instantaneously, rising to the surface, emerging as a cork above the surface of the water, and blowing away. This pro-eutectic solidification has nothing to suggest the solidifications with which we are familiar, and hence, though it is just as truly a solidification as that of the hypo-eutectic substance is, we may easily fail to recognize that it is.

CHAPTER 12

THE PHASE RULE[1]

309. In General.—What is the Phase Rule? What is it about, and why is it of importance to metallurgists? Today the chief motive for telling readers something about it is to put them on their guard against many of the current attempts to apply it. It is a most remarkable and valuable generalization; its conceptions help greatly toward getting a broad outlook on metallography; but its misconception has brought out a flood of obscuring writings. It tells us about the constitution toward which alloys tend, that which they reach when equilibrium is complete, when all tendencies have asserted themselves and have been complied with completely. But it tells us nothing directly about the intermediate stages through which those alloys pass in their attempt to reach that equilibrium, nothing about alloys which are out of equlibrium. And whatever it may teach us indirectly about such inequilibrium is so hedged about with the direct and indirect results of that inequlibrium that its application to any individual case is fraught with the greatest difficulty.

A hasty reading of the phase rule has led to a mass of attempts to apply it to conditions to which it is in its nature inapplicable, *e.g.*, to the constitution of hardened steel, which is evidently very far out of equilibrium. But this is like determining density with a telescope or area with a tuning fork; it is like the mad tea party at which the butter put into the works of the watch was the best butter. If the reader learns no more than to beware of attempts to decipher the conditions of inequilibrium by a law which touches only the conditions of exact equilibrium, if he rejects summarily all attempts to deduce the constitution of hardened steel by the application of the phase rule unless they show a complete mastery of the subject, he will have learned something of value. To which the busy reader may reply patly that if that is all he may as well stop at this point in the present chapter, and if he does I do not see that I can make any effective rejoinder, inclining to admit that if I were in his place I too might.

But those who nevertheless would learn something about this wonderful law by easy stages may well refresh their memory on certain elementary matters.

310. Subject Matter of the Phase Rule.—Just as crystalline rocks are composed of perfectly definite and distinct mineralogical entities called minerals, such as the mica, quartz, and feldspar of a granite, so our alloys when solid are in general composed of similar definite entities, which correspond to the minerals of our crystalline rock. These entities, these minerals in one

[1] *Cf.* Gibbs, *Trans. Conn. Acad.*, 1876, vol. 3, p. 152.

case and quasi-minerals in the other, are called "Phases," a term which will be more closely defined later. The phase rule has to do with these phases.

We have further seen that these microscopic constituents actually present in an alloy may or may not be those which should normally be present at the existing temperature. For instance we have seen that the constituents normal to slowly cooled steel are ferrite and cementite, partly free, partly mechanically mixed as the quasi-eutectic conglomerate called pearlite. We have further seen that, if the steel is cooled suddenly from above the critical range, instead of consisting when cold of these normal constituents, it actually consists chiefly of martensite. The martensite is not normal at this temperature; it is a transition form preserved by rapid cooling; when so preserved it is abnormal, and in this sense the rapid cooled steel is not in molecular equilibrium. The jurisdiction of the phase rule will be explained more fully after the rule itself has been set forth.

The phase rule is one which by higher analysis, and almost by inspection, helps us to answer such questions as (1) whether in any given alloy the constituents actually present are those which ought normally to be present, *i.e.*, whether the alloy is in molecular equilibrium in this sense; and (2) for any given transformation or change of state, what are the normal constituents which ought to result, *i.e.*, which correspond to equilibrium in this sense; and (3) many other related questions.

This it does essentially through telling us the "degree of liberty" of the alloy regarded as a system, *i.e.*, whether the existing constitution of the alloy has such a degree of stability that it can survive a change of temperature or of concentration, or cannot. Roughly speaking we may say that we find out what the "degree of liberty" is by comparing (1) the number of different "components" or "constituents," such as free metals (and metalloids if any), present in any given alloy and also certain other entities, with (2) the number of "phases," *i.e.*, the number of distinct, and as it were mineralogical entities or distinct "metarals," or quasi-minerals, present in it.

In general, for a given number of components, the fewer the phases present the greater the degree of stability.

311. Steel as an Illustration.—For instance, pure carbon steel contains two components, iron and carbon. If hardened by rapid cooling it may be taken provisionally to consist of three distinct metarals or phases, (1) the martensite which has been in large part preserved by the rapid cooling, unstable in the cold, tending to change back to ferrite and cementite; (2) a little ferrite and (3) a little cementite, both formed by the transformation of a little of the martensite in spite of the rapidity of the cooling. As a whole the molecular equilibrium is unstable; there is a residual tendency, and a powerful one though held in check, for the rest of the martensite to change back to ferrite and cementite. Here are two components, but those components are present in three distinct phases or conditions.

The ferrite and cementite of slowly cooled steel, on the other hand, are normal for the existing temperature and represent stable equilibrium. On

slightly re-heating slowly cooled steel, *i.e.*, annealed steel, its "system" undergoes no change, *i.e.*, it remains composed of ferrite and cementite. Here as before are two elements, iron and carbon, but these components exist in only two phases or conditions, ferrite and cementite. For our present purpose we may ignore the limited stability, the "meta-stability," of the cementite, its tendency to decompose under favorable conditions, into ferrite plus graphite.

The constitution or "system" of annealed steel, then, is evidently more stable than that of hardened steel, since it is not changed by this slight re-heating, this change of condition; or in other words it can without itself changing survive this change of environing condition. In the language of the phase rule the former system has a greater "degree of liberty" than the latter. All this will be defined more precisely later.

We know as a fact that the former system is more stable than the latter; but if it so happened that we were in ignorance of this, the phase rule would instruct us, without our having to make a direct experiment. And whereas we do happen to know it in the familiar case of steel, in many other like cases we should not know, unless we made a direct experiment, which of such two states was the more stable, and indeed we might not know anything about the stability, absolute or relative, of the two systems. Here, too, the phase rule would stand us in like stead.

In the comparison between rapidly and slowly cooled steel, the phase rule bases its finding on the comparison between the numbers (1) two components but in three phases in the former, and (2) still two components but in only two phases in the latter. This too will be explained more fully beyond.

312. The value of the phase rule to the metallurgist is chiefly as a guide for interpreting phenomena and planning investigations. When the investigator has before him a set of facts the explanation of which is not clear, he may on one hand proceed in ignorance, frame many conceivable theories by which these particular facts can be reconciled, and then proceed to test these different theories. The phase rule on the other hand may enable him to test them, and reject many or perhaps all of them immediately, by inspection and without making direct experiments; or it may immediately point out to him crucial tests which will enable him to decide quickly between two different theories.

Its present importance for students of alloys lies less in the present than in the prospective results of its application to this subject, in view of the present very active efforts of many investigators to apply it. This activity is so great that it would be hardly proper, even in such a work as the present, to pass the phase rule by without an attempt to outline its meaning.

Beyond this, a clear exposition of its applicability or jurisdiction is needed today to restrain the half initiated from misusing it and the uninitiated from being misled by these misuses. For instance, it is a favorite procedure of the half initiated to assert with complete confidence that such and such a

theory is untenable, because it offends the phase rule, though in fact that rule has nothing whatsoever to say concerning such theory. The phase rule defines the conditions of complete equilibrium; it may be likened to the laws of equity. But many of our theories have to do with conditions which are manifestly wholly out of equilibrium; the fact that they offend the phase rule is in general wholly irrelevant. Their conditions might be likened to martial law. As well might the laws of equity be invoked to explain or arrest the movements of the enemy in battle.

Let us now look into certain of these matters a little farther, especially into the conditions of equilibrium and the terminology of the phase rule, before taking up that remarkable rule itself.

313. Meaning of Equilibrium.—Aqueous solutions and molten alloys, such as salt or sodium nitrate in water, iron carbide in molten iron, etc., are in general homogeneous bodies. In solidifying they habitually split up into distinct entities, a solution of sodium nitrate splitting up during solidification into separate particles of ice and others of the salt, molten steel or cast iron splitting up into primary austenite or primary cementite, and eutectic austenite plus cementite. Further molecular rearrangements may occur after solidification is complete and even when the solid metal has cooled so far as to be very firm, *e.g.*, even at 450°, or for that matter at 66°. (See §168, footnote.) Thus in the yellow-hot or red-hot metal, cementite transforms into austenite plus graphite, austenite into martensite, and martensite into ferrite plus cementite, partly grouped together as pearlite.

The molecular rearrangements of both classes, but more especially the sluggish transformations in the solid metal, often remain very far from complete when the metal has completely cooled, (1) because cooling at any common rate may not give the time which their completion requires, and (2) because, thanks to the slowness of diffusion, two constituents which would react on each other were they in actual contact, may be isolated from each other by a third constituent. Were there sufficient time for diffusion, these two constituents thus isolated would reach each other, react, and bring the system to molecular equilibrium. Once the metal is cold, what we may call molecular rigidity or molecular viscosity prevents these transformations; the cold like a brake, as it were, locks the atoms in their existing condition, abnormal for this range of temperature and in that sense unstable, striving to change to the molecular grouping normal at the existing temperature, but restrained by this molecular rigidity.

Thus even in relatively slowly cooled alloys equilibrium may be very imperfect, solid solutions remaining far super-saturated and transformations being restrained. Of course rapid cooling, as in the hardening of steel, should leave the system still farther from equilibrium, so that there are strong residual unsatisfied tendencies to molecular re-grouping, re-arrangement, transformation.

If there are such unsatisfied tendencies pressing to assert themselves, such strained and eager springs, then on even slight re-heating and consequent

relaxation of the brake of molecular rigidity, the tendencies which have been resisted by the rapid cooling now re-assert themselves, the restrained transformations go on, the atoms whose sluggishness held them in the wrong path now repent and return, and a certain degree of re-grouping takes place.

314. An Example of Unstable Equilibrium.—Let the tempering of hardened steel illustrate this. Above the transformation range $A1$ to $A3$ the metal spontaneously assumes the condition of austenite normal to that upper range; if it is cooled slowly to below that critical range it changes spontaneously to the condition of ferrite and cementite normal to this lower range of temperature. If the steel is cooled rapidly from above the critical range to the room temperature, the change from austenite to ferrite and cementite is in large part arrested, so that most of the metal is caught in the intermediate condition of martensite even in the cold, although this condition is not normal to this lower range of temperature. This is the *hardening* of steel.

If this hardened steel is now slightly re-heated, even if this re-heating carries it only up to 66°C., part of the unstable abnormal martensite changes toward the condition of ferrite and cementite: the hardening is slightly mitigated, let down, or *tempered; i.e.*, the transformation which has been restrained by the rapid cooling now takes place to such degree as the slight re-heating and the slight molecular relaxation which the re-heating induces, permit.

Note particularly that the relaxation of the brake by such re-heating is very slight, so that only a fraction of the restrained transformation occurs on such re-heating.

Taking up the simile of §313,[1] there is the bent spring of the restrained transformation still striving to complete itself, but restrained by the brake of the low temperature applied by the rapid cooling before the transformation had time to complete itself. Relax the brake slightly by slight re-heating, and the spring can now begin to straighten itself; but in this very beginning of its own straightening, the spring loses a fraction of its tension, and this tension quickly falls to a point at which the present diminished grip of the brake suffices to master it, so that the spring moves no farther; metastable equilibrium between spring and brake is re-established, the restrained transformation goes no farther. Re-heat it still further, *i.e.*, relax the brake still more, the spring further straightens itself correspondingly, its tension falling correspondingly until it again becomes so light that the present friction of the brake again restrains it, and the transformation is again arrested.

Such is the state of affairs in a cold alloy if, regarded as a system, it is unstable.

315. Stable Equilibrium.—If, on the other hand, the several tendencies toward molecular re-grouping, toward transformation, have been fully complied with during slow cooling, so that the molecular constitution is

[1] Compare the Author, "Iron, Steel and Other Alloys," Boston, 1903, §§204 to 206, p. 230.

normal for the existing temperature, and thus in stable equilibrium; then if the alloy or other system is re-heated, no re-grouping or transformation takes place, for the simple reason that there is no residual unfulfilled tendency toward re-grouping, still ignoring the distinction between the limited or meta-stability of cementite, its tendency to transform into the truly stable system, ferrite plus graphite, and the complete stability of graphite.

316. Behavior on Re-heating a Criterion of Stability.—This, then, is one criterion of stable equilibrium; if such equilibrium exists, re-heating does not change the constitution; if not, it may. Conversely, if the constitution is changed by re-heating, it was not in equilibrium; if it is not changed by re-heating, it may have been.

This criterion must, of course, be applied with discretion and knowledge. For instance, if hardened steel is re-heated to 200°, so that a certain degree of tempering, of change of martensite toward ferrite and cementite occurs, thanks to the increased molecular relaxation at this temperature; if it is then cooled to the room-temperature, and then again heated to 100°, no further change will take place. But yet we know that the metal is still chiefly composed of martensite, *i.e.*, that the equilibrium is still unstable. The prior re-heating to 200° has allowed the molecular transformation to go so far that the residual tendency toward farther transformation, though it exists, is so slight that it cannot overcome the greater rigidity which exists at 100°. The fact that re-heating to a temperature lower than that to which the quickly cooled metal has previously been re-heated, induces no further change of constitution, is no evidence of stability of equilibrium.

317. Re-heating, as a Criterion of Stability, should Rise Past no Critical Point.—Conversely if steel is cooled so slowly that the transformation from austenite into pearlite with ferrite or cementite becomes complete, so that the cold metal is in stable molecular equilibrium, nevertheless its constitution will begin changing back to austenite when we re-heat it to a critical temperature such as *Ac*1 (approximately *PSK* in Fig. 23, p. 130). But this is no evidence that the system when cold was unstable, as the least reflection shows. Water at 80° is stable.[1] The fact that if heated to 100° or cooled to 0° it ceases to be stable and is transformed into steam or ice, is no evidence to the contrary. When we say that it is stable, we refer to stability within the range of temperature in which the existing temperature lies. This range of temperature is bounded by critical points, for instance 0° and 100° in case of water. The range may be narrow or it may be wide, and the existing temperature may be near or far from one or both boundaries of the range.

Thus if we are asking whether an existing system, as represented by some given alloy, is the system normal to the existing temperature and therefore in equilibrium, and if we seek to answer this question by noting whether on re-

[1] For simplicity of statement we here ignore the fact that the stability of the water requires that the atmosphere with which it is in contact must be saturated with water vapor for the existing temperature and pressure. In short, we ignore the gaseous phase for the moment. The water and its atmosphere evidently constitute a system, and if the atmosphere is thus saturated with water vapor the system is stable.

heating that system undergoes change or not—if we ask and seek this, then the re-heating which we perform (1) must be to a temperature higher than any which the alloy has reached since it entered its present range; and (2) it must not reach any critical point.

318. Reversibility and Lag.—This leads us to the subject of reversibility. The so-called "changes of state" gaseous to liquid, liquid to solid, which occur during cooling are naturally reversed during a subsequent heating; and the same is true of the transformations unaccompanied by such change of state which occur in cooling. For instance, if steam at atmospheric pressure is progressively cooled, it first condenses into water at 100°, and then this water freezes into ice at 0°; if we now raise the temperature, this ice will again melt at 0°, and the resultant water will change into steam at 100°. The changes of state in cooling are exactly reversed in heating, and they are reversed habitually at the very temperatures at which they occurred in cooling; freezing and melting both occur at 0°, boiling and condensation both at 100°.

If we have steel of exactly 0.90 per cent. of carbon and cool it from say 900°, on reaching a temperature of about 725° it changes from its initial condition of austenite into ferrite and cementite. If, after this transformation is complete, we re-heat the steel, it changes back into austenite. But the temperature, *Ar*1, at which this transformation occurs in cooling is considerably lower than *Ac*1, the temperature at which it is reversed in heating; the transformation is reversed, but at a higher temperature. The fact that there is such a difference of temperature is a clear indication that the transformation in cooling did not occur promptly on reaching the theoretical temperature at which it should begin, and that something of the nature of surfusion or undercooling took place during cooling. This inaccuracy of the reversing, then, is a first indication that during cooling the transformation lagged, and that during this time of lag the metal was in an abnormal state, one not of equilibrium.

If the changes which normally ought to occur in cooling actually do occur fully, so that the cooled alloy is in molecular equilibrium, then they are reversed fully on re-heating, so that the change which occurs in re-heating is exactly the reverse of the change which occurs in cooling, and quantitatively identical with it. If, however, the tendencies toward any given change are not fully complied with during cooling, so that the cold alloy is not in molecular equilibrium, then the changes which occur on re-heating are not the exact reverse of those which occurred in cooling, as will be explained in §320.

We have steel containing 0.90 per cent. of carbon; it has been cooled slowly; it consists of ferrite and cementite. We heat it past *Ac*1 and it changes into austenite; we cool it below *Ar*1 and it tends to change back to ferrite and cementite. There is thus the tendency on rising past the transformation-point to change from the normal low-temperature state (ferrite and cementite), to the normal high-temperature state (austenite); which change we may for brevity call the Low-to-High-Temperature-State Change, or the L-to-H change. Conversely, on cooling past the transformation-

point there is the tendency to change from the normal high-temperature state of austenite to the normal low-temperature state of ferrite and cementite, or for brevity the H-to-L change.

319. Special Technical Meaning of "Reversible."—A lagging change, one which occurs at a lower temperature in cooling than in heating, is often called "Irreversible," whereas a prompt one, one which occurs promptly at the equilibrium temperature in cooling, and because of its promptness occurs even the more promptly at this same equilibrium temperature in heating up, is called "Reversible." "Lagging," and "Lagless" are unequivocal, but they do not suggest this difference or lack of difference between the changes in cooling and in heating with which "Reversible" and "Irreversible" have now become associated. These latter names, therefore, though misleading, must be used in general.

320. Hardened Steel Illustrates Irreversibility.—Suppose that this steel is above the transformation-point $Ac1$, and therefore is austenite; suppose we cool it so quickly that, of the H-to-L change which tends to occur as soon as the temperature sinks below $Ar1$, only a first instalment, only n per cent. actually does occur, before the temperature has sunk too low to permit this change to proceed farther. Only n per cent. of this change has occurred; the remainder, the complement, $100-n$ per cent., is suppressed, *i.e.*, is arrested, but is still trying to occur. It is a spring restrained by the brake of low temperature.

We next relax the brake slightly by re-heating to 230° as in tempering (so that a straw-colored film of oxide forms), and a second installment of this suppressed complement, say a second n per cent., now occurs; so that all told $2n$ per cent. of the H-to-L transformation has now occurred, and the complement, $100-2n$ per cent. of it, still remains restrained by the brake. We then re-heat still farther as in tempering at 300°, so that the film of oxide on the steel turns blue, and we thereby relax the brake farther; thereon a third instalment, say an additional $4n$ per cent. of this H-to-L transformation occurs; so that now altogether $6n$ per cent. of it has occurred. There is still a suppressed or restrained complement of $100-6n$ per cent. striving to occur, but still prevented by the remaining grip of the brake. We next re-heat still farther say to 600°, and we thereby remove the brake apparently completely, so that the $100-6n$ per cent. complement which at 300° remained restrained, now occurs, and so that the whole of the H-to-L change takes place. The metal is annealed.

But if we finally carry the temperature a little higher, up above the transformation-point $Ac1$, the metal now changes from ferrite and cementite into austenite, in short the L-to-H change occurs.

Here, then, our actual transformations are as follows:

1. In rapid cooling (hardening) from above $Ar1$, only n per cent. of the H-to-L transformation occurs, and $100-n$ per cent. of it is restrained or suppressed.

2. In re-heating,

(A) The remaining $100-n$ per cent. of the H-to-L change occurs step by step as we reach first 230° (tempering straw), then 300° (tempering blue), and finally 600° (annealing).

(B) When the temperature rises above $Ac1$ the whole of the L-to-H transformation takes place.

Or in short, in cooling, n per cent. of the H-to-L transformation; in re-heating

(A) $100-n$ per cent. of the H-to-L transformation, followed later

(B) by 100 per cent. of the L-to-H transformation.

The transformations which here occur in heating clearly are very far from the exact reverse of those which occurred in cooling.

321. Why Irreversibility Implies Absence of Equilibrium.—The reasons have really made themselves clear in this example. If cooling is too rapid to permit the H-to-L transformation to perfect itself, then the transformations which actually occur in such cooling are but an initial instalment of the normal ones, the complement of that instalment being suppressed by molecular rigidity. In order that the transformations which occur on re-heating such rapidly cooled material should be exactly the reverse of those which have occurred in such a cooling, they should consist in simply reversing this initial instalment. But instead of this, what first occurs on re-heating is the remainder, *i.e.*, suppressed complement of the H-to-L transformation, because heating relaxes the molecular rigidity which has suppressed or restrained that complement; and when the heating eventually reaches the transformation temperature, it is not an instalment but the whole of the H-to-L transformation which is now reversed, the initial n per cent. instalment plus the suppressed complement.

Could the heating leap back absolutely instantaneously to the transformation temperature without consuming any time, at that temperature it would be only the initial instalment of H-to-L transformation which would be reversed, so that in this case the transformation in re-heating would be the exact equivalent of that which occurred in the sudden cooling; or in other words there would be complete reversibility, although the rapidly cooled steel was certainly not in equilibrium. But such an instantaneous rise can never occur; it is indeed hardly thinkable; this rise from the cold to the transformation temperature must occupy a finite time, no matter how brief. Our heating must traverse the range between the room temperature and that transformation temperature, and what takes place in that range is the suppressed complement of the H-to-L transformation, and not simply a fraction of the reverse or L-to-H transformation.

Thus one criterion of whether the transformations normal to cooling have been completed, *i.e.*, of whether the cold alloy is in a state of molecular equilibrium, is that the series of transformations which it has undergone in cooling are exactly reversed in heating, or in short are *reversible*.

Thus, to say that the condition or molecular constitution of an alloy or other system reached through cooling is in molecular stability, and to say

that the transformations through which it has passed are reversible, are synonymous. To say that the transformation temperatures during heating (*i.e.*, during rise of temperature), are higher than those during cooling is to say that there is lag and a period of molecular instability; to say that the transformation temperatures in heating and cooling are identical is to say that there is no such lag, and that the transformations in both heating and cooling occur promptly on reaching their normal temperature.

322. Terminology of the Phase Rule. Component and Phase.—The meaning of these two words must be grasped clearly. Let me define them, and explain them by means of examples.

In the terminology of the phase rule, the "components" of an alloy or other system are the free elements and those compounds which in the nature of the case are undecomposable under the conditions contemplated, and so play the part of elements; those which enter directly into the equilibrium of the system. They are the uncombined elements and compounds equivalent to elements in this sense. Lead and tin (elements) are the components of a lead-tin alloy, iron and carbon (elements) are those of steel, chloride of sodium and water (chemical compounds) are those of a salt-water solution.

The components may pass from one physical state (solid, liquid, or gaseous) to another (water boils or freezes), and they may combine with each other and dissolve in each other in the form of a solution or compound (iron and carbon uniting to form austenite or cementite), or emerging thence into the free state (cementite splitting up into graphite and iron); and when they so merge, each of the compounds or solutions thereby formed tends to become homogeneous. It must become homogeneous before equilibrium can exist. Every such homogeneous compound or solution and every free element is a phase; and each of them is a distinct and different phase for every different physical state. Phases then are "the homogeneous states, whether of freedom, solution, or combination, and whether solid, liquid, or gaseous, into which the components present pass or group themselves." The components are the entities in play, the entities of which we are studying the reciprocal behavior; the phases are the transitory stages, states, or conditions, physical and chemical, through which the components pass as they are heated up and cooled down, or as their pressure rises and falls. As temperature falls and pressure rises, hydrogen and oxygen show themselves in six phases, comparable in a way with the phases of the moon, (1) and (2) as dissociated gases, (3) as steam, (4) as water, (5) as common ice, and (6) as Tamman's denser ice. By homogeneousness I here mean only chemical homogeneousness, such as exists in a crystalline mass, which however homogeneous chemically is not homogeneous physically, but anisotropic (§333).

The usual definitions are, for phase "a mass chemically and physically homogeneous or a mass of uniform concentration, the number of phases in a system being the number of different homogeneous masses, or the number of masses of different concentrations:"[1] and for components "the substances of

[1] W. D. Bancroft, "The Phase Rule," Ithaca, p. 1, 1897.

independently variable concentration in the phase or system under consideration."[1] These definitions seem very exact unless in requiring physical homogeneousness, but they may not convey a very clear idea to an unwilling mind; good as tests, but not as explanations.

323. A. Examples of Component and Phase. *The Components are Elements.*—Applying this to our metals and alloys, pure molten iron consists of a single component, iron, and a single phase, molten iron. If it cools to the freezing point and begins freezing, different parts of the metal are simultaneously in two different phases, (1) molten iron and (2) solid iron; these are the two phases of the single component, iron. There is, strictly speaking, a third phase, the vapor of iron and usually the atmospheric oxygen and nitrogen; but we may simplify our whole discussion by ignoring the vapor in all cases, and suiting our formulas to this intentional omission.

In any iron carbon alloy there are two components, iron and carbon. If such an alloy of eutectic ratio, 4.30 per cent. carbon, is molten, there is one phase, this molten solution; when it begins to solidify and break up into the eutectic of plates of austenite interstratified with those of cementite, three phases are present: (1) the molten alloy, a molten solution and therefore a single as it were mineralogical entity, (2) the plates of solid austenite and (3) the plates of solid cementite.

Again, in a pure steel consisting solely of iron and carbon, these two unchanging permanent elements are the two components, while the transient decomposable compound, cementite, and the equally transient solution, austenite, are phases, and so are the gaseous, molten, and solid states. As the temperature rises past the critical or transformation range, past the melting point, past the boiling point, past the dissociation point, and then descends again, these two unchanging components, iron and carbon, enter into reciprocal play. They start as two phases, free iron, ferrite, and iron carbide, cementite; on passing the transformation range they pass from these phases into the solid-solution phase of austenite; when raised to the melting point the austenite phase shifts into the phase of molten carburized iron (molten austenite, may we call it?). On reaching the boiling point this phase shifts (let us assume, for example) to that of gaseous carburized iron; at the still higher dissociation point this phase (let us again assume) shifts into that of free gaseous iron and free carbon. On re-combining, re-condensing, re-freezing, and re-transformation, these phases are passed through in the reverse order. Ferrite and cementite, solid austenite, molten austenite (?), gaseous austenite (?), gaseous free iron and free carbon, these, let us assume, are the successive phases (some of them hypothetical), through which the unchanging components, iron and carbon, pass.

Here is a value of this word phase; it keeps us in face with the fact that we are regarding these systems as the resultants of some change from some other system; we regard a solid granite as resulting from state of fusion or at least from metamorphosis; we regard the constitution of a solid alloy habitu-

[1] *Loc. cit.*

ally as resulting from the solidification of that alloy from its prior molten state.

324. B. Examples in which the Components are Chemical Compounds.—We have thus far considered systems of which the components are elements; let us go a step farther and consider some in which the components are chemical compounds, playing the part of elements in that they are undecomposable under the conditions assumed.

A sodium-nitrate water solution is frozen, and this ice is again melted; here the two components, sodium nitrate and water, start in the phase or condition of liquid solution; they then in freezing pass into the phases of solid sodium nitrate and solid water or ice; on re-melting they pass back into the phase of liquid solution. Here the phases are three: (1) the liquid salt-water solution, (2) solid nitrate, and (3) ice. But what are the components? Manifestly the water and the nitrate; it is they that, each remaining undecomposed, play back and forth between these different phases, quite as the uncombined elements iron and carbon did in the case which we have been studying. In the present case it is clearly not the nitrogen, the sodium, the hydrogen, and the oxygen which are to be regarded as components. They do not directly enter into play; each is initially combined with one or two elements, the nitrogen with the sodium and part of the oxygen, and it is no longer free to act as nitrogen. It is not directly between these elements, but between the water and the salt that equilibrium exists; we are studying not the reciprocal behavior of these elements but of these compounds. Therefore these compounds, nitrate and water, are the components.

Or, to look at it in the language of the usual definition, the salt and water are "of independently variable concentration in the phase or system under consideration" in that we can vary the concentration of either of these independently of the other. We can add or remove water (by evaporation), and so shift the degree of concentration; but we cannot remove nitrogen without removing the sodium and oxygen with which it is combined, nor can we remove hydrogen without removing the oxygen with which it is combined. No one of these four elements can be removed from the system or added to it independently of the element or elements with which it is combined, without changing the system into a wholly new and different one. To remove some of the sodium would change the system into a four component one, consisting of water, nitrate, and the free nitrogen pentoxide and oxygen left by the removal of that sodium.

Again, in a typical granite composed of muscovite (potash) mica, quartz, orthoclase (potash) feldspar, we have (1) the three phases, each of these minerals being a phase, and (2) three components, silica, alumina and potash. It is true that there are four elements, silicon, aluminum, potassium, and oxygen present, assuming that each of our minerals is quite pure. But for purposes of equilibrium we need not consider the number of elements; it is quite sufficient that, ignoring them, we concentrate our attention on the number of phases and of components, because in the nature of the case, in

melting and re-freezing our granite, in passing back and forth through different ranges of temperature, though these three components, these three oxides, may group themselves together in different ways, no one of them would be decomposed; as far as equilibrium is concerned they might as well be elements.

Again, when hydrated cupric sulphate is dissolved in water, then again crystallized out, then dehydrated, then again re-hydrated and re-dissolved, the two unchanging entities are anhydrous cupric sulphate and water; it is these two that are the components; it is they that play back and forth through the phases of anhydrous salt, solid hydrated salt, and solution of different degrees of concentration. It is true that there are three chemical elements, copper, sulphur, and oxygen; but this is beside the mark, for neither the cupric sulphate nor the water is liable to be decomposed through the changes of temperature contemplated. No one of these three elements enters directly and by itself into any reaction or change of state or of concentration, or into the equilibrium of the system; nor can either be varied independently of the others. That is to say, we cannot remove from the system part of its oxygen without simultaneously removing either the hydrogen or the copper and sulphur with which that oxygen is combined.

Allotropic Modifications are Distinct Phases.—Again, every distinct allotropic form of an element is a distinct phase. If graphite, diamond, and amorphous carbon coexisted, they would constitute three distinct phases, because a mixture of different allotropic forms, and indeed of crystals of different forms, is not "chemically and physically homogeneous."

325. Composite Eutectics are not Phases.—Although cold steel of 0.90 per cent. of carbon, when slowly cooled, consists solely of pearlite, nevertheless it consists of two phases, viz., the ferrite and cementite of which that pearlite is composed. As pearlite is not a mineralogical but a lithological entity, not homogeneous but heterogeneous, not a single substance but a mechanical mixture of two, it is not a phase at all, for a mechanical mixture does not constitute a phase. So although hypo-eutectoid steel when slowly cooled consists of pearlite plus ferrite, it contains only two phases, ferrite and cementite; for the fact that part of this ferrite is mechanically intermixed with the whole of the cementite in the form of pearlite does not affect the number of phases present.

Homogeneousness of Phases.—It has been said that every phase must be homogeneous; let us look at this matter further. Oil floating on water, or mixed up with the water as an emulsion, typifies a di-phase system. A homogeneous solution of salt in water is a mono-phase system. But how about a heterogeneous solution, either liquid, molten, or solid? Is it a phase, or if not, what is it?

We may answer this in two ways. First, the phase rule bears upon the subject of equilibrium; but it is perfectly evident that no system can be in equilibrium which contains a heterogeneous solution, because there must exist in that solution an unfulfilled tendency to diffuse so as to efface that

heterogeneousness. Thus it is not necessary to resort to the phase rule to decide as to the equilibrium of such a system.

Second, in a heterogeneous solution there are in a sense an infinite number of phases, because every different degree of concentration represents a different phase. But, as we shall see, if we insist upon applying the phase rule to such a system, that rule shows it to be unstable.

Physical Actions.—In a broad discussion of this subject it would be necessary to consider variations in both temperature and pressure. For our present purpose the whole discussion may be simplified by considering the pressure as constant, and the temperature alone as changing, and this will be done in the remainder of this chapter.

Independently Varying Conditions.—In considering the equilibrium of a system, and the influence of changes of conditions on that equilibrium, there are two of these varying conditions which we must have in mind: varying temperature, and varying degree of concentration of any solutions which are present, whether those solutions are solid on one hand, or liquid or molten on the other hand.

Arbitrary and Conformable Variations in Condition.—Let us recognize that, when temperature and concentration vary, they may vary conformably with each other, or arbitrarily, *i.e.*, without regard to each other.

For example, in a beaker stands a solution of salt in water, at 20°, and just saturated for that temperature. We cool the mass to 10°, and the solution thereby becomes supersaturated, because the water has less solvent power for salt at 10° than at 20°; accordingly the excess of salt crystallizes out, and a new equilibrium is thereby reached. But this has destroyed the mono-phase system, salt solution, and substituted for it a di-phase system, salt solution plus solid salt. This variation in temperature was arbitrary; there was nothing done to compensate for it.

Suppose, however, that when we cooled the mass to 10° we had simultaneously poured in enough pure water, also at 10°, to dilute the liquor to the saturation-point for 10°; no salt would have crystallized out in this case, because we varied the second variable, concentration, conformably with the first, temperature. The system has survived the simultaneous changes in the two variables. In the first case the variation of temperature was arbitrary, but in the second instead of an arbitrary variation we had conformable variations of the two variables.

But if on cooling to 10° we had poured in half the quantity of water which we have just supposed, salt would still have crystallized out, because this smaller quantity of water would not suffice to prevent the supersaturation which the cooling tends to cause; in this case the variations in the two variables would be arbitrary instead of conformable.

326. Degree of Liberty. The Phase Rule.—By the degree of liberty of a system we indicate whether that system can, without tending to break up and become converted into some other system, survive an arbitrary variation in one of these two variables at a time, temperature and concentra-

tion; or in both of them simultaneously; or in neither of them. We are purposely ignoring variations in pressure, and also ignoring the vapor present, to simplify our discussion.

The degree of liberty is measured by the terms invariant or nonvariant, monovariant, and divariant.

An invariant or nonvariant system is one which cannot survive a change either of temperature or concentration (*i.e.*, of either of our variable conditions), but breaks up and passes into some other system if either is changed, and this breaking up, due to a change of either of these conditions, cannot be prevented by any conformable change in the other condition. In other words, it is a system which can be in equilibrium only at one certain temperature, and with one certain degree of concentration of each phase present.

A monovariant system is one which can survive a change of either temperature or concentration if accompanied by the conformable change in the other, but cannot survive simultaneous arbitrary changes in both temperature and concentration.

A divariant system is one which can survive simultaneous arbitrary changes in both temperature and concentration.

The degree of liberty of an invariant system is 0, that of a monovariant system 1, and of a divariant system 2.

We learn the degree of liberty of any given system, or in other words we learn whether it is invariant, monovariant, or divariant, by the formula

$$L = n + 1 - r,$$

in which

L = the degree of liberty { 0 for an invariant system, 1 for a monovariant system, 2 for a divariant system,

n = the number of components present, and

r = the number of phases present.

This formula is *the phase rule.*[1]

A cold alloy if invariant is clearly unstable, since heating breaks it up;

[1] This formula for the phase rule is based upon our purposely ignoring not only the gaseous phase always present, but also the variations in pressure and other physical actions, surface tension, magnetism, etc. In a more general discussion these ought to be taken into account; and if we recognize variations in pressure, the formula becomes

$$L = n + p - r,$$

in which p = the number of so-called physical actions, temperature and pressure. Here, as before, L is 0 for an invariant, 1 for a monovariant, and 2 for a divariant system; or, as is sometimes said:

a nonvariant system has $n + 2$ phases
a monovariant system has $n + 1$ phases
a divariant system has n phases.

In any given case the formula given in this footnote, and that given above in the text, lead to the same result. Thus, a mixture of ice and water is nonvariant by either formula; that it is invariant by the formula of the text $L = n + 1 - r$ is shown in §249; that it is also nonvariant by the wider formula of this footnote is easily seen. The number of constituents is 1, water; the number of phases is 3, ice, water, and steam—*i.e.*, the vapor of water floating above the mixed ice and water; the number of physical actions is 2, temperature and pressure. Hence, $L = 1 + 2 - 3 = 0$.

a cold monovariant alloy, on the other hand, is stable, and our stable alloys are characteristically of this type, though they pass through an invariant stage in freezing. This confirms the assertion already made that for a given number of constituents the degree of liberty, *i.e.*, the stability, is the greater the smaller the number of phases present.

327. Jurisdiction of the Phase Rule.—Unfortunately a great deal of confusion has arisen from hasty applications of the phase rule by those who fail to recognize the limits of its jurisdiction. It tells us of the conditions of equilibrium, the conditions toward which every system tends. But these tendencies are not always obeyed; the phase rule cannot always enforce itself.

Recognize distinctly that it does not tell whether a system actually will break up on changing the temperature, but whether it will tend to break up. A hardened high-carbon steel or cast iron, though it has only two components, carbon and iron, may contain graphite, austenite, ferrite and cementite, which are surely phases, and also martensite and osmondite, of which the former is generally believed to be a distinct phase and the latter may be. If we ignore the martensite and osmondite, the degree of liberty is $2 + 1 - 4 = -1$. If we include martensite, it is $2 + 1 - 5 = -2$, and if we include osmondite it is $2 + 1 - 6 = -3$; so that the degree of liberty is a minus quantity. The phase rule does not say that such systems cannot arise, but only that they are out of equilibrium at every temperature, and that they are at all times striving to reach equilibrium through the suppression of enough phases to leave the system invariant.

So, too, of a nonvariant system which is in equilibrium at only one temperature the phase rule does not say that it cannot exist at any other temperature; but that at such other temperatures it is out of equilibrium, and is always striving against the existing passive resistance or lag to reach equilibrium by the suppression of one phase, as for instance by its conversion into one or more of the others. For instance, the nonvariant system graphite + ferrite + cementite can be in equilibrium at only one temperature; yet we find it throughout a wide range of temperature. The phase rule does not deny this, but only says that this system is out of equilibrium at all but one of these temperatures and is striving against lag to reach equilibrium, to become monovariant, by suppression of the cementite through the reaction $Fe_3C = 3Fe + Gr$.

It further says that the natural presumption is that, if at constant temperature, or if within the limits of a certain range of temperature which includes no transformation point, *e.g.*, Region 4 of Fig. 23, §165, any change which occurs in a system thus out of equilibrium is a change toward equilibrium through the suppression of some phase, and that the burden of proof lies heavy on those who hold that the change occurs in the opposite direction, away from equilibrium and with an increase in the number of phases.

It further says that whatever change occurs in any system must be a change toward and not away from equilibrium, even though that change may actually work through temporary increase in the number of phases

present and thus on superficial examination may seem to be away from equilibrium. For instance, a steel with 1.60 per cent. of carbon if heated to 1,100° consists of austenite; if it is now cooled to 900° a certain quantity of free cementite separates, beginning to form when the temperature sinks past ES. If it is held long enough at 900° this cementite begins transforming into graphite, and while this transformation is going on the system is invariant, with two components, iron and carbon, and three phases, austenite, cementite, and graphite. Here is a change away from a monovariant to an invariant system, and hence away from equilibrium; a change, moreover, which seems to occur at any temperature in this range, 725° to 1,135°.

The phase rule does not assert that such a transformation cannot occur, but that every such change must be toward equilibrium, and hence that this transformation must be tending to complete itself by complete suppression of the cementite and its conversion into austenite plus graphite. Further, that, because this change must be away from the less and toward the greater equilibrium, cementite must be less stable than graphite, and the initial system, austenite plus cementite, less stable than the final one toward which the transformation is working, austenite plus graphite.

328. Examples to Test the Phase Rule.—Some examples which we will now take up will serve both to explain and to test the phase rule. Let us in each case first find what degree of liberty the phase rule assigns to the system under consideration; *i.e.*, whether the phase rule shows it to be nonvariant, monovariant or divariant; and if for instance the phase rule shows it to be monovariant, let us next examine the conditions of the case to see whether, judged from these, it really seems to be monovariant. Let us in short test the truth of the phase rule by an examination of several different systems.

For ease of reference, the elements, phases, components and degree of liberty of each of these cases are put together in condensed form in Table 17.

Pure Water in the Act of Freezing.—Here water and ice are both present, and each of them is a distinct phase; hence the equation is

Degree of Liberty = number of constituents + 1 − number of phases.

$$L = 1 + 1 - 2 = 0$$

so that the system is invariant.

Let us test this invariance. First, no change in composition is possible, since the chemical composition of the two phases is the same; the diffusion from one phase, the water, into the other phase, ice, would leave the composition pure water, *i.e.*, unaltered. Looking at it in another way, any change in composition would be a change from this system to another system; if the substance added dissolved in the water, the system would be new, because a solution would be substituted for pure water. If it did not dissolve but remained apart, it would form an additional distinct and separate phase, and thus change the system from a di- to a tri-phase one.

The only change possible then is one in temperature. If we heat the

TABLE 17.—CONDENSED STATEMENT OF THE EXAMPLES TAKEN TO TEST THE PHASE RULE, §§328 TO 330 INCLUSIVE

System	Elements	Components	Phases	Entities present which are not phases	Degree of liberty
Freezing water.....	H, O	Water	Water Ice	.....	Invariant
Solidifying iron.....	Fe	Fe	Molten iron solid iron	.....	Invariant
Sodium-nitrate solution above the freezing point	Na, N H, O	$NaNO_3$ H_2O	Sodium-nitrate solution (liquid solution)	.....	Divariant
Sodium-nitrate ice.....	Na, N H, O	$NaNO_3$ H_2O	Sodium nitrate Pure ice	Eutectic of salt plus ice	Monovariant
Sodium-nitrate solution during selective freezing	Na, N H, O	$NaNO_3$ H_2O	Ice Mother-liquor	.....	Monovariant
Ditto during eutectic freezing	Na, N H, O	$NaNO_3$ H_2O	Ice Solid sodium nitrate Mother-liquor	Eutectic of ice plus sodium nitrate	Nonvariant
Molten steel of 0.4 per cent. carbon	Fe, C	Fe, C	Molten solution	.....	Divariant
Ditto while solidifying	Fe, C	Fe, C	(1) molten iron (2) primary austenite	.....	Monovariant
Ditto at 1,300°, solid..	Fe, C	Fe, C	Solid solution (austenite)	.....	Divariant
Ditto at 730°, Region 7 (Fig. 23)	Fe, C	Fe, C	(1) austenite (2) ferrite	.....	Monovariant
Ditto at 725° j^V, the recalescence	Fe, C	Fe, C	(1) austenite (2) ferrite (3) cementite	Pearlite (eutectoid of ferrite plus cementite)	Invariant
Ditto at the room temperature when slowly cooled = annealed	Fe, C	Fe, C	(1) ferrite (2) cementite	Pearlite	Monovariant
Ditto at the room temperature, after quenching from above *Ar*1	Fe, C	Fe, C	(1) austenite (2) martensite (3) ferrite (4) cementite	Ditto in small quantity (uncoagulated as troostite)	Unstable
Gray cast iron at the room temperature	Fe, C	Fe, C	(1) ferrite (2) cementite (3) graphite	Pearlite	Invariant
Chilled white cast iron	Fe, C	Fe, C	(1) ferrite (2) cementite (3) graphite (4) martensite	Pearlite, uncoagulated	Unstable

system, some of the ice will melt, and the temperature will remain constant until the whole of it has melted; and during this constancy of temperature the system will remain unchanged, since a change in the proportions of the two different members of a system is not reckoned a change in the system;

the phase rule is qualitative. But before we can actually raise the temperature above the melting point, the whole of the ice must melt, *i.e.*, any actual elevation of temperature breaks up the system by completely removing one of its phases, ice, turning it from a di-phase into a mono-phase system.

In like manner if heat is abstracted, *e.g.*, by setting the vessel which contains the freezing water in liquid air, the temperature will remain constant until the water has all frozen, *i.e.*, no change in temperature will occur until the system breaks up by the complete removal of one of its phases, the water, and so changes from a di-phase into a mono-phase system. In short, the system breaks up whenever the single variable condition, temperature, varies. It is therefore invariant; so that the phase rule proves to be true in this case.

Pure iron in the act of freezing in like manner contains one constituent, in two phases, molten and solid iron; so that $L = 1 + 1 - 2 = 0$, *i.e.*, the system is invariant, exactly as in the last case. No change in temperature can occur until either the whole of the solid iron or the whole of the molten iron has been eliminated by changing into the other phase (all the solid iron melting or all the molten iron solidifying), the system thereby changing from di-phase into mono-phase. The phase rule thus proves to be true in this case also.

Sodium Nitrate Solutions above the Freezing Point.—There are two constituents, nitrate and water; there is one phase, a liquid solution; $L = 2 + 1 - 1 = 2$, the system is divariant. Now let us test this divariance.

Let us assume that the degree of concentration of the solution and the temperature are jointly represented by point D, 10 per cent. of sodium nitrate, Fig. 20, §132. Clearly the liquid solution will remain a liquid solution, *i.e.*, the system will remain mono-phase and unchanged if we lower the temperature arbitrarily to f without any change in composition, *e.g.*, by setting the vessel in liquid air; or if without changing the temperature we change the composition to f' by pouring in water; or if we simultaneously both dilute and cool the solution and bring its temperature and composition to f''. In short, the system remains mono-phase, *i.e.*, it is not broken up, even if both variable conditions, temperature and degree of concentration, are varied arbitrarily and each without reference to the simultaneous variation in the other. The system is therefore divariant.

Sodium-nitrate-ice.—Assuming for simplicity that nitrate and frozen water or ice are absolutely insoluble in each other, the nitrate-ice, say of composition and temperature e, is a conglomerate of (1) the excess of pure ice above the eutectic ratio of 63.1 per cent. water, 36.9 per cent. nitrate and (2) the eutectic of interstratified pure nitrate and pure ice. Here then are two components (nitrate and water), and two phases, (1) pure ice and (2) pure nitrate. $L = 2 + 1 - 2 = 1$, the system is monovariant. As before let us test this monovariance.

Here we need not enquire as to whether the system could survive a change in concentration simply because such change is unthinkable, on our present assumption that solid nitrate and ice are reciprocally absolutely insoluble.

Hence the only variation to be considered is temperature. Clearly if the conglomerate of crystals of ice and other crystals of nitrate are heated (provided of course that the temperature does not reach the melting point of the eutectic), no change will occur. The system is not split up on undergoing this variation, *i.e.*, it is monovariant.

Sodium Nitrate Solutions in the Act of Selective Freezing are Monovariant.—Still considering this same 10 per cent. nitrate solution, let us suppose that it has cooled to − 6°, *h*, Fig. 20, §132, so that in cooling from *g* to *h* selective freezing has been going on, the mother-liquor has been generating and depositing flakes of ice, and has thus enriched itself so that it now contains 15 per cent. of the salt, and therefore is represented by point *b*, because at every instant during selective freezing the mother-liquor is of the salt content for which the existing temperature is the freezing-point, or in other words the mother-liquor has been sliding down *AB*.

Here are two components, water and nitrate, and two phases, (1) ice, and (2) solution or mother-liquor containing 15 per cent. of the salt dissolved in 85 per cent. of water.

Test of Monovariance by Changing the Temperature.—The equation is $L = 2 + 1 - 2 = 1$. The system is monovariant. Let us test its monovariance.

The composition of the ice cannot vary, since we have for simplicity assumed that nitrate and ice are reciprocally completely insoluble; but the concentration of the mother-liquor can vary.

First let us see what happens if the temperature is lowered 6° to point *N*. Some more ice will freeze out, the mother-liquor will be enriched up to 27 per cent. of the salt (point *Q*), but the phases will be the same as before, ice and mother-liquor; the system in short has not been broken up, for the change in concentration of a solution is not to be reckoned as a change in the system as such. In that the solution has this new degree of concentration it might indeed be called a new phase; but it is for our purpose the equivalent of the old one which it replaces.

But how about this same change in concentration; have not two conditions really varied, both temperature and composition, without breaking up the system; and is not the system therefore divariant in spite of the phase rule? Certainly not, for the change which has taken place in composition is simply that conformable with or corresponding to the change in temperature, and is not an arbitrary change. There has thus been but one arbitrary change.

Indeed it is this spontaneous shifting of concentration which has kept the system from breaking up and changing from (1) ice plus (2) mother-liquor, into the system (1) ice plus (2) solid salt, as we see on the slightest consideration. For, if by any device it could have been possible to prevent the change of concentration, so that when the temperature reached − 12° the mother-liquor still contained only 15 per cent. of the salt as it did at − 6°, that mother-liquor would have frozen, because the new temperature, − 12°

is 6° below the freezing point (− 6°) of such a solution. A failure of the composition to shift conformably with the temperature would in effect be an arbitrary variation of the composition.

Test of Monovariance by Changing the Concentration.—Thus far we have considered changes of temperature, and we have found the system monovariant, failing to break up if the temperature is changed, provided of course that the concentration shifts conformably, but breaking up if concentration does not so shift, *i.e.*, in effect shifts arbitrarily. Let us next take up the parallel case of changes in concentration, starting as before at − 6°, at *h*. Let us now shift the salt content of the mother metal to 27 per cent., *S*, by pouring into it enough solution still richer in the salt and also at − 6° to effect this change. If we make simultaneously the corresponding change in temperature, *i.e.*, if we lower the temperature from *S* to *Q*, from − 6° to − 12° by external cooling, our system will remain unchanged. This is in effect what we did in the last case.

But if we do not lower the temperature, it is evident that our ice will begin melting and entering the mother-liquor, because, so long as this does not decrease the salt content of that mother-liquor to below 15 per cent., *b*, that mother-liquor will remain liquid. We have seen (§336) that this is the general condition which determines whether a solid shall melt and enter the liquid with which it is in contact. If the new mixture which such entry would create would remain liquid, whether as an aqueous solution or as a molten alloy, *i.e.*, if it would be above its own freezing point, the solid will so melt and enter that liquid. This same reasoning shows that this re-melting of the ice will continue till it dilutes the mother-liquor down to 15 per cent. Then the re-melting will stop, for the simple reason that any further dilution of the mother-liquor by the melting of that ice would bring its salt content to the left of *b*, *i.e.*, would create a mixture which at this temperature would be below its own melting point.

But suppose we now try again, and pour off this liquor, and add a new lot of 27 per cent. nitrate solution; this will in the same way dissolve more of our ice, and will keep dissolving it until the salt content of the liquor shall have fallen to 15 per cent. So our steps to keep in contact with our ice at this temperature, a liquid with more than 15 per cent. of the salt, do not at first succeed; manifestly no matter how many times we replace the liquor with new, its salt content will always sink down to 15 per cent. by taking up some of the original ice, until the whole of the ice shall thus have been melted away; in short until our system shall have been broken up by the vanishing of one of its members, the ice.

Clearly the same result would follow if, instead of pouring off our mother-liquor and replacing it by a new lot, we were to enrich it in the salt, at constant temperature, − 6°, by pouring into it a lot of liquor containing say 20 per cent. nitrate, initially at this same temperature, − 6°.

Having failed to make our system survive at − 6° with a liquor richer in nitrate than 15 per cent., it is evident that we cannot make it survive with

a liquor poorer than 15 per cent., because the existing temperature, − 6°, is below the freezing point of any such poorer solution, and hence such a solution cannot exist at − 6°. If we started with such a poorer solution at a higher temperature, above its freezing point, and cooled it to − 6°, it would freeze on the way down and enrich itself up to 15 per cent. by the time it reached − 6°, so that we should still have the same old system, the initial ice plus the ice thus frozen out, forming together a single phase, and our 15 per cent. solution for the other phase.

If there were some way in which we could keep extracting the salt from our liquor, and so forcibly bring its nitrate content below 15 per cent., as fast as we did this it would begin freezing because the existing temperature, − 6°, is below the freezing point of any such poorer liquor. What would freeze out would be pure ice, and this would continue freezing out till its departure enriched the liquor up to 15 per cent., when this freezing would cease, because the existing temperature, − 6°, is above the freezing point of any richer liquor. If we could withdraw still more of the salt, more ice would freeze out, and this might continue till at last we had withdrawn all the salt, when the last of the water would freeze, and our original di-phase system would be reduced to a mono-phase one, ice.

Summary of these Tests.—In short, the system (1) ice and (2) nitrate solution, cannot survive any change in its concentration, whether by enriching or by impoverishing it in the salt, unless that change is accompanied by a corresponding shifting of the temperature. An enriching in nitrate would reduce it to a mono-phase system, the solution itself, by melting all the ice; and an impoverishment in salt by an imaginary process of removing the salt step by step, would reduce it to a mono-phase system, ice.

This result is wholly in accordance with the phase rule, since a shifting of concentration unaccompanied by the corresponding shifting of temperature is in effect an arbitrary shifting of both concentration and temperature, which, as the phase rule teaches, a monovariant system cannot survive.

Thus in every respect our system satisfies our tests of its monovariance.

The Same Sodium-nitrate Solution at the Freezing Point of the Eutectic, − 18.5°, *B*, Fig. 16.—Let us suppose that the whole of the excess of water has frozen out, so that the mother-liquor has been enriched to its eutectic ratio, B, of 36.9 per cent. nitrate and 63.1 per cent. water; and that part of this eutectic mother-metal has frozen, while part still remains unfrozen.

The phases will be (1) ice, including (A) the excess over the eutectic ratio, frozen out between *g* and *l*, and (B) the plates of ice contained in the already frozen part of the eutectic; plus (2) solid nitrate in the frozen part of the eutectic; plus (3) the still unfrozen part of the eutectic mother-liquor. Three phases, and as before two constituents, water and nitrate.

The equation becomes $L = 2 + 1 - 3 = 0$, *i.e.*, the system is invariant. Let us as before test this invariance. The possible variations are (1) rise and (2) fall of temperature; (3) enrichment and (4) impoverishment in nitrate of the still unfrozen mother-liquor. Let us test these variations

separately; if the system is truly invariant, any one of them should break it up.

Rise of Temperature.—If we apply heat, the now frozen part of the eutectic will begin to melt, and the temperature will remain constant until the whole of it has melted; then only will the temperature begin to rise. But this complete melting of the eutectic will remove the whole of our solid nitrate. In short a rise of temperature can occur only after the system has broken up through the elimination of this second phase, solid nitrate; or in other words the system cannot survive a rise of temperature.

Fall of Temperature.—In like manner, if we abstract heat by external cooling, the still liquid part of the eutectic mother-liquor will begin freezing, and the temperature will remain constant until the whole of it has frozen, *i.e.*, until the liquid phase of the system has been eliminated; only then will the temperature fall below its initial point of − 18.5°. In short, a fall of temperature can occur only after the system has broken up through the elimination of its third phase, the mother-liquor; or in other words the system cannot survive a fall of temperature.

We saw that in the range of selective freezing, at *h*, Fig. 20 (as distinguished from the present eutectic freezing), either a rise or fall of temperature could be survived provided that the concentration of the mother-liquor shifted correspondingly. But here the system could not be saved by any such device, for any rise of temperature would re-melt the now solid part of the eutectic, thus eliminating our second phase, solid nitrate, no matter what the composition of the adjacent mother-liquor. Further, any fall of temperature would eliminate our third phase, the mother-liquor, no matter what its composition; because no solution of nitrate in water can remain unfrozen below our initial temperature of − 18.5°, for it is the freezing point of the most fusible solution of the series. Thus any fall of temperature must substitute solid nitrate and ice for the mother-liquor; but this would break up our system, changing it from a tri-phase to a di-phase one, consisting of (1) ice and (2) solid salt, because every solution of nitrate in water in freezing must turn into a mixture of particles of pure ice and others of pure nitrate, thanks to the assumed complete insolubility of these two substances in each other when solid.

Here again, no shifting of concentration can enable our system to survive any change of temperature, for any rise of temperature must eliminate the solid nitrate phase, and any fall of temperature the mother-liquor phase.

Change of Concentration.—(3) Enrichment and (4) impoverishment of the molten mother-liquor in nitrate. As the temperature is now by assumption at the freezing point of the eutectic, and as this is the most fusible solution, the temperature is below the freezing point of every other solution. Consequently it would be impossible to substitute for this mother-liquor any other, either richer or poorer in nitrate, because any other such liquor would freeze at this temperature. Any liquor either richer or poorer thus substituted would at this temperature be solid; and because solid it would be

a mechanical mixture of plates of solid nitrate and other plates of ice, these two substances having been assumed for simplicity to be absolutely insoluble in each other when solid. But our system would then be (1) ice, including both (A) the excess of water over the eutectic ratio frozen out between *g* and *l*, and (B) the ice in the new solid alloy now added; plus (2) the solidified salt in this newly added solution. In short it would be a new and di-phase system, instead of our original tri-phase system, for the third of our original phases, the mother-liquor, would have been eliminated. In short our system cannot survive either enrichment or impoverishment of the molten phase in tin.

We have thus found that the system cannot survive any of the four possible changes in temperature and concentration; so that it satisfies all our tests of its complete nonvariance.

329. The Iron-carbon Compounds.—To take a simple series of cases let us select steel containing 0.40 per cent. of carbon, jj^{VI} of Fig. 23, §165, at six different temperatures, 2,000° (*j*), 1,440° (between j^{I} and j^{II}), 1,200°, 730°, 725° (j^{V}) and 0° (j^{VI}), representing (1) the molten condition, (2) the state of selective freezing, (3) the austenite state, (4) the generation of ferrite just above the recalescence, (5) the recalescence, and (6) the common or room-temperature state. This latter case we will subdivide into two, those of (A) hardened (rapidly cooled) and (B) unhardened (slowly cooled) steel respectively.

In every case there will be two constituents, carbon and iron; the system will be invariant if there are three phases, monovariant if there are two, divariant if there is only one.

1. At 2,000°, *j*, there is one phase, molten carburized iron; the system is divariant.

2. At 1,440° during selective freezing, there are two phases, molten carburized iron and solid primary austenite. The system is monovariant.

3. At 1,200° there is only one phase, austenite. The system is divariant.

4. At 730° there are two phases, (1) the free iron which has already been generated as ferrite by and within the plastic redhot austenite, below j^{III} and (2) the residual austenite which has generated and still encloses it. The system is monovariant.

5. At 725°, j^{V}, the recalescence, there are three phases, ferrite and cementite (partly interstratified in the form of pearlite if the cooling is slow enough, uncoagulated as sorbite if it is less slow), which have already resulted from the transformation of the austenite, and the residual austenite which has not yet split up into ferrite and cementite. The system is invariant.

6A. At the room-temperature, j^{VI}, if the transformation and decomposition of the austenite have been complete, so that only pearlite with ferrite as the excess substance remains, there are only two phases, ferrite and cementite; the system is monovariant; it is in stable equilibrium; it is not affected by re-heating, *e.g.*, to 300°.

6B. At the room-temperature j^{VI}, if the steel has been hardened by a cool-

ing so rapid that part of the austenite has failed to decompose into ferrite and cementite, then there are at least three if not four phases present.

The true number of phases is not yet known certainly, because the nature of the martensite, of which the rapidly cooled steel consists chiefly, is not known. It is probable that after any rapid cooling the transformation has in some groups of molecules gone so far that some ferrite and cementite have formed, and that in other groups it has been so far restrained that some traces of austenite persist though none of these end components can be identified under the microscope. If so, austenite, martensite, ferrite, and cementite are simultaneously present. If martensite is a distinct phase, four phases are present, and the degree of liberty is $2 + 1 - 4 = -1$. It is a minus quantity. If martensite is not a distinct phase, there are three phases present, and the degree of liberty is $2 + 1 - 3 = 0$. In either case the system is not in equilibrium. When re-heated, *e.g.*, to 200° as in tempering, part of the martensite and of the austenite if any is present, change into ferrite and cementite, though so finely divided that they cannot be recognized. The uncoagulated transition stage, troostite, appears.

The reasoning already given in §328 suffices to explain these several cases.

330. The Phase Rule Applies only to Systems Properly So-called.— The unthinking might be puzzled by trying to apply the phase rule to cases to which it is manifestly inapplicable, to cases which are not systems at all in the sense of the phase rule.

Thus some might reason as follows: "I place a lump of cold sulphur upon a lump of cold iron. Two components, sulphur and iron; two phases, sulphur and iron. Degree of liberty $= n + 1 - r = 2 + 1 - 2 = 1$; hence this is a monovariant system; hence if sulphur and iron are heated they will not react upon each other according to the phase rule, unless they rise past some critical point properly so-called.

So with a mixture of oxygen with either hydrogen, iron, sodium, or any other readily oxidizable element. In each case there are two components and two phases, a monovariant system and hence one which according to the phase rule should not change with a change of temperature unless this passes some critical point. But in fact I know that the components of each mixture do react, and I see no reason to believe that this reacting requires passing some critical point properly so-called, for I happen to know that iron finely enough divided is pyrophoric, and will combine with oxygen at the common temperature. Hence the phase rule is wrong and misleading.

The answer is that such mixtures are not systems at all in the meaning of the phase rule. The uncombined mixture of hydrogen and oxygen is not a system of two phases into which water has spontaneously dissociated itself at the existing or room-temperature.

Let the case of the ferrite and cementite illustrate this. While these substances in slowly cooled steel, resulting from the decomposition of the austenite of which the steel consisted when it was above the transformation

temperature, form a system, which is di-phase; yet if we were to isolate certain particles of this pure cementite and certain other particles of pure ferrite and shake them up together while cold, this would not constitute a system. The phase rule has nothing to say of such a mass.

Thus if it were true that water spontaneously dissociated itself into a two-phase mixture of hydrogen and oxygen at room-temperature, we should justly infer that this system was monovariant, *i.e.*, that they would not recombine on raising the temperature, provided of course that such rise does not pass any critical point.

331. The Phase Rule in one Aspect is Qualitative, not Quantitative.—While it is true that the phase rule is quantitative as regards the number of components and phases, yet it has nothing to do with the quantity of any of those components or phases, and in this sense is qualitative. Let the transformation of a hypo-eutectoid steel of 0.40 per cent. of carbon illustrate this. When in cooling to just below the line *GOS* (*Ar*3) in Fig. 23, a small quantity of ferrite has separated out in the form of an extremely fine net-work around the grains of austenite, as shown in Fig. I of Plate 8, the two phases are:

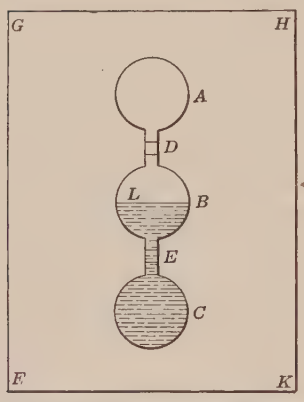


FIG. 33.—Equilibrium is not quantitative but qualitative.

1. Pro-eutectoid ferrite, as a network, forming, say 1 per cent. of the whole, and

2. Still untransformed austenite, forming the meshes which that net-work encloses, forming say 99 per cent. of the whole.

As the transformation progresses, the ferrite thus generated forms a continuously increasing proportion of the whole, and the remaining untransformed austenite a continuously decreasing percentage of the whole; yet the two remain always in equilibrium, and remain one and the same system.

Indeed, the least reflection shows that equilibrium—and it is with equilibrium that the phase rule has to do—is a purely qualitative thing, and has no relation to quantity. Let a simple example verify this idea.

In a large chamber *FGHK* (Fig. 33) is placed a steam boiler *B*, with which are connected a mud-drum *C* and a steam-drum *A*. The boiler is filled with water to about level *L*, and, an escape valve being open from *A* into the open air, the temperature of *FGHK* is raised to 100°C. (212°F.) The water boils until all air is expelled. Then this exit valve is closed, so that no steam can escape and no water can be fed in, in short so that the system *A*, *B*, *C* is isolated from without; but the valves *D* and *E* are open, so that the boiler and its drums are in free communication with each other. The temperature is raised to say 121.4°C. (250.5°F.), and the water in *B* evaporates until the pressure of steam (its vapor tension) in the upper part of *B* and in *A* reaches 2 atmospheres (29.5 lb. per square inch). At this pressure and temperature the steam and water are in equilibrium with

each other. The further evaporation of water is just balanced by the simultaneous condensation of steam, so that to the eye neither evaporation nor condensation appears to take place.

This equilibrium has nothing to do with the quantity of water and steam, or with the proportion between the two, as is readily seen. Assume that, temperature and pressure remaining constant, we close valve E; manifestly this will not induce either further evaporation or further condensation: it will have no effect whatsoever. Suppose that now, holding this valve closed, we disconnect this mud-drum C and remove it altogether to the outside and destroy it. The quantity of water remaining in the system will be only about one-third as great as originally, while the quantity of steam remains unchanged; the ratio between the quantity of water and of steam has changed greatly, but this change has not in the least affected the equilibrium.

The same would be true if, closing valve D, we had removed the steam in the steam-drum A from the system; and also if, holding the temperature and pressure constant, we had replaced and re-connected the steam or the mud-drum, or both of them, reopening either or both valves D and E. Thus we see that no matter what change we bring about in the quantities of water and steam, this change does not affect the equilibrium, provided that temperature and pressure remain constant.

332. Invariant Changes Represented by Horizontal Lines.—By that I mean that the temperature of any given change which passes through an invariant system is constant and independent of the proportion of the several phases present.[1] This follows directly from the phase rule and the law that equilibrium is qualitative and independent of the relative proportion of the various phases present. The fact that the system passed through is invariant implies that it can be passed through at only one temperature, because it is stable at only one temperature, and hence, under conditions of equilibrium can come into existence at only one temperature. That this temperature is constant and independent of the proportions between the various phases follows from the fact that equilibrium is independent of those proportions.

This law is illustrated by the horizontality of every eutectic freezing line, and of every eutectoid transformation line, in equilibrium. For instance, in the selective freezing of a sodium nitrate water solution, the temperature gradually falls till the mother-liquor has been brought by the selection to the eutectic salt content, 36.9 per cent., when selection ceases, and the residual mother-liquor freezes at the freezing point of that eutectic $-$ 18.5°. But because the eutectic is simply that solution the salt content of which gives it the lowest freezing point, and because selection will bring the mother-liquor to this salt content no matter what its initial content, the composition of the eutectic is constant; and because its composition is constant its freezing point also must be constant, and independent of the initial salt content. Hence this transformation from the monovariant system mother-

[1] Compare Roozeboom, *Zeit. phys. Chemie*, 1900, vol. 34, p. 451.

liquor + ice or mother-liquor + nitrate, through the invariant system mother-liquor + ice + nitrate, to the final and monovariant system ice + nitrate, occurs at constant temperature, − 18.5°.

Lag may of course deflect this line from the true horizontal. Indeed transformations which are extremely tardy, like the graphitizing of cementite, Fe_3C + austenite → Gr + austenite, involving the simultaneous presence of three phases, and hence invariance, may actually be spread out over a very wide range of temperature, and we may be wholly unable to determine the single temperature at which, in equilibrium, they would occur.

CHAPTER 13

EVIDENCE THAT THE STRUCTURE OF METALS IS CRYSTALLINE

Part I. Evidence Obtained without Deformation

333. The Crystalline and the Amorphous.—Invaluable as our usual classification of matter into solid, liquid, and gaseous is, a cross classification such as follows, into the crystalline and the amorphous, is even more important from a philosophical point of view:[1]

State	Crystalline (geometrically anisotropic)	Amorphous (isotropic)
Gaseous		Gases.
Liquid	Liquid crystals	Probably most liquids.
Solid	Crystalline solids	Amorphous or vitreous solids (undercooled liquids).

Isotropic or amorphous means having properties identical in all directions, as in the case of gases, of most liquids, and of uncrystalline solids such as glass, varnish, clay, and putty. These transmit and resist pressure equally in all directions, and they have no cleavage.

Anisotropic as distinguished from isotropic means having properties which vary with the direction. Slate and wood are weaker across than along the cleavage, grain, or fiber, and therein are anisotropic.

Crystals and crystalline bodies are not only anisotropic but geometrically anisotropic. Their cleavages are arranged symmetrically about fixed axes, which in turn have fixed relative directions. This is not true of the cleavages of slate or the fiber of wood.[2] Hence we must distinguish between bodies which, like slate or wood, are anisotropic and those which, like crystals, are geometrically anisotropic.

The essence of the crystalline state is first that the molecules group themselves into minute crystal units which may be likened to the bricks of a house or the tiles of a mosaic pavement; and second that these adjoining units lie parallel and in symmetry, or in short that they have identical direction or orientation, as soldiers are arranged in platoons, companies, and regiments.

[1] This classification is slightly modified from those of Tammann ("Krystallisieren und Schmelzen," Barth, Leipzig, 1903, p. 5) and of Desch ("Report on Diffusion in Solids," from *Rept. British Assoc.*, 1912, p. 3).

[2] See the Author, "What is the Essence of Crystalhood?" *Metallographist*, 1902, vol. 5, p. 52.

This identity of orientation, of internal geometrical arrangement, extends throughout each of the crystalline grains of which we see the polygonal sections in our micrographs, but naturally changes at the boundary line of each grain to a different orientation in each of the adjoining grains. It is to this geometrical internal structure that all the characteristics of the crystalline state are due.

Metals, iron included, are primarily crystalline, though certain parts of any given metallic mass may exist in the amorphous state. That they are thus primarily crystalline we shall see from evidence collected in the remaining chapters of this volume, and that certain parts of them may be amorphous we shall see in Chapter 20. Incidentally we shall have to outline the general conditions of some of this evidence, such as the mechanism of the deformation, certain features of which are evidence of this crystalhood.

334. The Evidence that the Metals are Crystalline.—Speaking broadly, they have all the characteristics by means of which we recognize the crystalline condition in other bodies, save that of modifying polarized light in transmitting it. This their opacity prevents. But even here they show their crystalhood by their specific action in reflecting polarized light and, at least in the case of copper, in reflecting X-rays.[1] This evidence may conveniently be divided into three classes, as follows.

(A) The Evidence Reached without Deformation

1. That the metals, including iron, form true idiomorphic crystals under favorable conditions.

2. That they readily form dendritic crystals during solidification.

3. That they dispose extraneous solid matter geometrically about certain of their own planes.

4. Closely related to 3, that their etching pits and gas bubbles are arranged geometrically.

5. That the brightness of an etched microsection is uniform throughout each grain, but varies from grain to grain, and further that this variation itself varies both in degree and in sign with the direction of the light.

6. That their eutectics and eutectoids have the structure typical of those of other crystalline bodies, such as rock masses and the crystalline deposits from aqueous and other solutions.

(B) That Reached by Means of Plastic Deformation without Rupture

7. That pressure on a polished surface develops geometrical patterns.

8. That the deformation planes in metals are arranged geometrically.

9. That many metals readily form twin crystals.

[1] Desch, "The Hardness of Solid Solutions," *Trans. Faraday Soc.*, May, 1915, vol. 10; Bragg, *Phil. Mag.*, VI, 1914, vol. 28, p. 355; *Engineering*, 1915, vol. 99, p. 369; Bragg, W. H., and W. L., "X-rays and Crystal Structure," Macmillan, New York, 1915.

(C) That Reached by Rupture

10. That they yield crystalline fractures.

11. That they can be cleft into geometrical solids.

335. Crystalhood no Bar to Ductility.—While a great development of the crystalline organization makes for brittleness by establishing continuous and well-developed cleavage planes, which are necessarily planes of weakness, yet the fact that the crystalline structure of so ductile a metal as lead is extremely well marked and prominent shows that the crystalline state need not exclude ductility.

336. Space Lattice.—The molecules which compose a crystal are thought to be not in actual contact with each other, but separated by very minute yet perfectly regular distances. Let us conceive a system of equidistant vertical poles arranged in longitudinal and transverse rows, the transverse rows being at right angles to the longitudinal ones, like those of an apple orchard. Let us next imagine that along the ground runs a system of wires, one wire passing through each longitudinal row of poles and one through each transverse row. Above this first system of wires run other like systems, the distance between these layers of wires being the same as that between rows of poles.

Each layer of wires divides the horizontal plane running through it into squares, and the wires vertically over each other, taken jointly with a given row of poles, divide the vertical plane running through those poles into vertical squares. These horizontal and vertical squares jointly divide the whole space into a series of cubes, which may be likened to a mass of equal-sized cubic crates packed as closely together as possible. Now imagine a molecule set at every intersection of wires with poles, and the wires and poles then evaporated, leaving the molecules suspended in space. Just as the crates divided the whole space up into cubes, so do imaginary planes drawn through the centers of these molecules. Each of these cubes is a "space lattice."

This conception can be varied greatly. The distances between the longitudinal rows of poles, though all alike, may differ from those between the transverse rows, or from those between the layers of wires, or all three sets of distances may differ. Moreover, the poles need not be vertical, but may all slant longitudinally of the system, or transversely of the system, or both longitudinally and transversely. Moreover, instead of a single molecule there may be a group of molecules at each corner of each cube. These groups, though alike in any one system, may vary from system to system. Instead of being at the corners of the cubes, they may be at the centers of the cube faces, and so forth.

In any given crystal "all the space lattices are geometrically identical or are characterized by the same elementary parallelopipedon."[1]

[1] Von Groth, *British Assoc. Report*, 1904, "Encyclopædia Britannica," 11th Edition, 1910, vol. VII, p. 585.

Such a space lattice owes its coherence to the attraction between its molecules, and this attraction varies from point to point and from direction to direction. The attraction and hence the properties along a plane running through the four molecules of one face of the cube differ from those along a parallel plane running midway between molecules, and these in turn differ from those along an inclined plane, and these differ with the inclination of the plane. Hence we should not be surprised at the existence of planes of low cohesion, nor at the coexistence of several sets of such planes, say some cubic and some octahedral. Nor should gliding or twinning planes which are not cleavage planes surprise us, nor slip planes which are neither.

It is hardly necessary for us to go further, and consider the "point systems," or groupings of interpenetrating space lattices, or the interpenetration of these point systems to make up the crystalline mass. Here "point" refers to the center of gravity of a molecule or group of molecules.

337. A crystal unit is the smallest mass that is compatible with the retention of this crystalline structure. It may be of an order of magnitude much larger than that of the molecules themselves, or even than that of a space lattice, though Bragg's observations on the reflection of the Röntgen rays point to an order of magnitude less than that of the molecules.

338. Cleavages are planes of low shearing strength, normal to the direction of minimum tensile strength. They are invariably geometrically arranged, that is they are planes with rational indices. Moreover, they are strictly symmetrical. Thus if any one crystal face has cleavage planes parallel to it, all the other like faces belonging to the same crystalline form have cleavage planes parallel to them and of the same ease of cleaving. The distance between parallel cleavage planes of any given set represents the space lattice of the crystal itself, so that it is of molecular, or rather of crystal unit magnitude. Thus any given cleavage is an essential and integral part of the crystalline organization in which it occurs.

339. Partings are like cleavages in a general way, in that they are planes of low cohesion, usually parallel, but differ from them in occurring only at intervals, which, though they may be either regular or irregular, great or relatively small, yet are of a magnitude greater than that of crystal units. Partings may be caused by the presence of foreign matter, such as sheets of cementite in austenite or ferrite, or of thin layers of ferrite in austenite, and they may occur at the boundaries of twin lamellæ (Chapter 22). Thus a parting differs from a cleavage in being not an essential part of the crystalline organization, but a weakness developed by extraneous causes along certain geometrical planes of the crystal or other body. It is rather accidental or extrinsic than intrinsic. The planes along which slate cleaves, for instance, are not true cleavages but partings.

Partings can sometimes be distinguished from cleavages by their being confined to the specimens from certain localities, and sometimes by their occurring at detectably great distances apart in any given specimen.

340. "Slip planes" and "gliding planes" are terms often used indiscriminately to represent not only planes along which true slip occurs without twinning but also those along which twinning occurs without slip. For clearness I shall speak of the former as slip planes and the latter as gliding planes, which are by definition twinning planes also.

341. Motion Planes.—These four kinds of planes may be called collectively "motion planes," because they lend themselves to one or another kind of motion, cleaving in the case of cleavages and partings, twinning along the gliding planes, and slip along the slip planes.

342. The idiomorphism of metallic crystals has already been illustrated by the crystals of austenite and of cementite shown in the frontispiece, growing out into cavities formed in the solidifying mass by the gradual ebb of the molten tide on which they feed.

Iron forms on sublimation very beautiful cubes, with octahedral and other modifications of the isometric (cubic) system. Osmond and Cartaud[1] found that the habit of the sublimed crystals was identical for alpha and beta iron, that is for all temperatures below $A3$ or about 900°, cubes modified at the edges but never at the apices, but that for gamma iron, that is for temperatures above $A3$, the cubes, which were often modified by octahedra, were frequently combined along truncations of their apices, and had no modifications along their edges. These differences of habit are not in themselves evidence of allotropy, for they are of an order common in other mineral species, and are referable to the differences in the genetic conditions.

The apparent increase in the tendency toward the octahedral form with rise of temperature tallies with the habitual octahedral form of the crystals which form from the molten, both in steel and cast iron.

Cubes of iron formed below $A3$, by reducing vaporized ferrous chloride by means of vapor of zinc or of sodium, are shown in Fig. A of Plate 15, and dendrites formed above $A3$ built up of hollow cubes are shown in Fig. B of the same plate. Here each of the terminal cubes has a re-entering angle corresponding to that formed by the grouping of the cubes below.

The wonderful crystals found by Grignon in cavities in cast iron are very probably of austenite. Rudimentary idiomorphic crystals are very often found coating the interiors of cavities, especially the central axial cavity or pipe, formed during the solidification of steel. Even the upper surface of a solidifying ingot may be covered with well-marked if rather rudimentary idiomorphic crystals.[2]

[1] *Annales des Mines, Memoires*, 1900, vol. 18 p. 113, and Figs. 1, Plate II, and 18, Plate III. Also *The Metallographist*, 1901, vol. 4, p. 119, and Figs. 2 and 31. They prepared their crystals by sublimation, reducing iron from its volatilized chloride by means of hydrogen, following Péligot's method, or of volatilized zinc after the method of Poumarède, or of volatilized sodium.

[2] Belaiew shows beautiful octahedra on the surface of steel ingots of 1.80 and 2.30 of carbon. "Crystallisation, Etc., of Steel," *Journ. Russian Technical Assoc.*, 1909, Figs. 4 to 7; *Rev. Metallurgie Mem.*, 1912, vol. 9, p. 325. The furrows which very often line the sides of the pipe in steel ingots are evidently of crystalline origin in part, though their genesis is difficult to establish. The Author, *Trans Amer. Inst. Min. Eng.*, 1907, vol. 38, p. 40; *Journ. Iron and Steel Inst.*, 1907, No. I, vol. 73, p. 520.

But apart from these unusually favorable conditions of free growth within a molten mass or by sublimation, idiomorphic crystals completely enclosed in solid metal and not near or related to any cavities are very common. The great straight parallel-sided white bands in Figs. G, Plate 1 and N, Plate 3 are evidently the cross-sections of idiomorphic cementite crystals, probably unterminated, and so are the white rhomb and rhomboid in Fig. D of Plate 15.[1]

The curved flakes which graphite habitually forms, and the flake-mimicking colonies of graphite particles about ferrite skeletons in gray cast iron, may be regarded as idiomorphic crystals in a sense. In taking on this flake form the graphite is not simply occupying a space which it finds vacant, but is arranging itself according to its own natural crystalline habit, in a form little related to the environment. In order to assemble in their present places the particles of graphite have had to displace the particles of solid austenite which lay there. Such flakes are seen in Figs. E and F of Plate 12, and E of Plate 13.

The good idiomorphic crystals in which the native metals and alloys often occur suffice in themselves to establish the crystalline nature of metals. Native crystals of gold, silver, copper and amalgam are sometimes very well formed. Native idiomorphic crystals of many other metals occur, for instance, platinum, palladium, iridosmine, and arquerite, a silver amalgam.

The artificial idiomorphic crystals of bismuth are of extraordinary perfection and beauty.

343. Dendritic crystals[2] form very readily on the solidification of metallic masses. Indeed the great crystals of austenite shown in the frontispiece, lengthened by "parallel growth," are in effect dendrites unusually wide for their length, and the columnar crystals in the outer parts of steel ingots are only dendrites of a special form.

The typical long narrow dendritic crystals, which suggest tree forms, and thus have given rise to the name "dendrite," differ from those in the frontispiece only in having a ratio of length to thickness similar to that common among trees. Dendrites in a non-metallic mineral mass are shown in Fig. E of Plate 15.

The dendritic structure is very common in white cast iron, as shown in Figs. J and K of Plate 3, and A, B, and C of Plate 6. Appropriate etching

[1] This figure seems to represent a specimen of cast iron partly purified by Saniter's process, and cooled slowly. Stead, *Journ. Iron and Steel Inst.*, 1898, vol. 53, No. I, p. 182, and Fig. 25, Plate XXI, after p. 176.

[2] A dendrite or dendritic crystal is a tree-like aggregation of a great number of similarly oriented crystals, each of which is usually very minute. A dendrite differs from a common idiomorphic crystal only in the general outer shape into which its component minute crystals are aggregated. The trunks and branches of dendrites are often somewhat curved. The curved patterns which the frost works on the window pane are thin dendrites.

See for instance Heycock and Neville, "On the Constitution of the Copper-tin Series of Alloys," *Phil. Trans. Royal Soc.*, 1904, A, vol. 202, Figs. 1, 2, 10, 15, 16, 20, 42, 47, 49, 75, and 76 following p. 69 of reprint for interesting dendrites in copper-tin alloys.

methods may develop in steel, even after it has been rolled, a degree of arborescence much greater than the mere presence of columnar crystals in the outer parts of steel ingots suggests. Fig. C of Plate 15 shows how the dendritic ingot structure has persisted, not only during transformation, but also through the mechanical processes of shaping the ingot into this rolled wheel.

This structure is more persistent in phosphoric than in pure steel, because the concentration of the phosphorus during solidification, as a filling between the branches of the pine-tree crystals, is effaced only very slowly by diffusion. It will be considered further in §770 in connection with the ferrite ghosts to which it so often gives rise.

344. Bas-relief dendrites of extraordinary beauty and perfection often form on the upper surface of metals which solidify under favorable conditions. For instance the upper surface of the flat ingots of commercial antimony habitually shows these beautiful reliefs, like that of Campbell[1] shown in Fig. C of Plate 16.

In particular Humfrey[2] has recorded the sequence of steps in the growth of such bas-relief dendrites rising above the surface of lead while solidifying in a flat dish. First, a main trunk shoots out from the side of the dish across the surface of the lead, somewhat as ice crystals shoot across the surface of a thin body of water. Then, at points spaced along this trunk, main branches shoot out at right angles, and from these in turn minor branches at right angles, and so on. As we shall see in §371, he was able to identify the etching pits and the slip bands which he developed in this mass as definitely and geometrically related to the geometrical structure of this dendrite.

That these dendrites stand thus in relief means that, as this solidification and crystallization proceed, the level of the still unfrozen surface sinks progressively relatively to the already frozen and relatively rigid dendrites. This comes about in three coöperating ways. The main trunk which first forms is attached firmly to the sides of the mould or other vessel. The mould becomes heated by the metal and expands. The vertical component of this expansion raises the dendrite slightly above its initial level, while the horizontal components of the expansion, increasing the width and breadth of the vessel, lower the level of the remaining molten. At the same time the metal, both that which has already solidified along the bottom of the mould and that which is still molten, continues to cool and hence to contract, and thus adds to the lowering of the remaining molten surface. Thus the dendrite is gradually emerging above the level of the molten, as it grows by the addition of new metal where its margin is still wetted by the remaining molten. Hence the combination of growth by means of new particles deposited out of the molten along the then existing edges of the dendrite, and hence deposited

[1] Campbell shows many bas-relief dendrites and other crystalline markings of extraordinary beauty in aluminum, cadmium, platinum, silver, tin, and zinc, in his "Effects of Straining and Annealing in Aluminium, etc.," *Appendix IV, Sixth Report Alloys Research Committee, Excerpt Proc. Inst. of Mechan. Engin.*, May–June, 1904.

[2] *Phil. Trans. Royal Soc.*, 1903, A, vol. 200, p. 225.

while free to obey the directive crystalline force, with a lifting of these newly deposited layers out of the molten as they deposit, so that their crystalline form is not later obscured.

If any given minute area on the very margin of the already emerged solid dendrite is still coated on its upper surface by a thin film of the tide above which it has thus just emerged, this film in solidifying may crystallize conformably with the solid branch or twig on which it is then reposing, so that the upper surface of the dendrite remains very sharp; or, if it adopts only imperfectly the orientation of the branch on which it dries, it at most softens slightly the sharpness of the crystalline angles and summits of that branch.

345. Each Grain or Cell of a Pure Metal a Crystal.—Any common metallic mass, even when it lacks or is irrelated to the dendritic structure, is in a sense as truly crystalline as the rare idiomorphic crystals, with this difference, that instead of being made up of single idiomorphic crystals it consists, like a loaf of sugar or a block of crystalline marble, of an aggregation of little crystalline grains or cells, which can be separated from each other under favorable circumstances. Such isolated grains of ferrite are seen in Fig. A of Plate 40, with their characteristic smooth but strongly curved faces. Like grain faces can be seen in what is called a "*granular*" fracture of steel. It is the polygonal section of these grains that we see in the microsections of pure metals, such as Figs. A, Plate 3, and T, Plate 9.

These grains are roughly like the straight-sided cells within a mass of soapy froth in an early stage of blowing soap bubbles. Save that the outer surface of the outer bubbles is rounded, each bubble of the froth is a fairly regular polyhedral cell with very straight walls. In an imaginary section through such a froth, the cell walls between bubbles correspond to the boundaries between grains in the microsection of a simple metal, such as the ferrite of low-carbon steel in Plates 23 and 25, of wrought iron, Fig. F of Plate 14, and of malleable castings, Figs. A and B of Plate 4; or of an alloy consisting of a single constituent, like the austenite of the invar of Figs. C and D of Plate 30; or of the austenite of which carbon steel consists when above the transformation range, Fig. B of Plate 16;[1] or of the ferrite, both pro-eutectoid and pearlitic, of such a steel when below that range, as in Figs. A of Plate 3, T of Plate 9, and E of Plate 16.

In metals which have been treated properly the individual grains are usually very small or even microscopic, but at temperatures high enough to give mobility these grains tend to increase in size, apparently by the absorption of the smaller and weaker by the larger and stronger (see footnote to §705). Indeed our processes are in part directed to making them small. But by favoring the grain growth, the grains of silicon steel may be made as much as 2 in. across, as shown in Fig. A of Plate 16.[2] Meteorites, indeed,

[1] Wark, *Metallurgie*, 1911, vol. 8, after p. 734, Fig. 727.

[2] This figure shows a strip of silicon steel kindly given me by Mr. W. E. Ruder. The greater part of it is made up of three enormous grains, each occupying the full width of the strip, 2 in.

often consist of a single grain, because their constituents have almost unlimited opportunity for coalescence during their nearly infinitely slow solidification and cooling, whether as parts of a sun, planet, asteroid, or satellite later broken up by impact with another stellar body, or as a small fragment ejected from a vast mass still molten. Here the cooling which accompanies their recession from the molten mother body must be extremely slow, for it is only that caused by the inconceivably slow lessening of the angle which the mother body subtends, and hence of the heat intake, to replace their loss of heat into the absolute vacuum at the absolute zero through which they pass.

The shape of these grains is almost identical in the different pure metals.

Each of these polyhedral grains, in spite of lacking the outer form of a crystal, is a true crystal as regards its internal structure, having by definition uniform orientation throughout, with uniformity of cleavage and all the other consequences of this uniformity of orientation. Note, for instance, how the slip bands of Fig. D of Plate 22 and the Neumann bands of Figs. C and D of Plate 31 are nearly straight and parallel within each grain, but change direction sharply at the grain boundaries. The slip bands usually stop short on reaching a grain boundary, but those which cross over into another grain usually change direction at the boundary as in Fig. B of Plate 38.

As we shall see in §419 this straightness does not always hold true of the slip bands, especially of those in ferrite, whether because we include two different things under this one class, or because they do not always represent crystalline movement, or for other reasons. So, too, the shape of the etching figures changes from grain to grain (§371).

Throughout each grain, at least in the case of copper not in the cast state, and of soft steel, the orientation is uniform from boundary to boundary.[1] These boundaries are likely to be something more than mathematical surfaces. In the first place foreign matter is likely to be massed here in the intergranular spaces, and in the second place it is reasonably surmised that a discontinuous amorphous sheet separates each grain from its neighbors.

This material is coarsened by high heating, at least to 1,020° (in this case for 2 hours to 1,300° in hydrogen), after undergoing plastic deformation, which in this case consisted apparently in sheet rolling. See also Ruder, *Trans. Amer. Inst. Min. Eng.*, 1913, vol. 47, p. 569, and *Journ. Industrial and Engineering Chemistry*, 1913, vol. 5, p. 452. The greatest coarseness is had when the stretching is from 2½ to 3 per cent. This stretching is usually done in the cold, but in the case of some thin sheets, 0.014 in. thick, which failed to coarsen after cold stretching, a great coarseness was induced by stretching from 8 to 12 per cent. at 900° to 1,000°, instead of in the cold. W. E. Ruder, private communications Nov. 4 and 9, 1914.

[1] Heyn, *Journ. Iron and Steel Inst.*, 1904, No. I, p. 378. Also "Mikroskopische Untersuchungen an tief geätzten Eisenschliffen," Berlin, 1898. This agrees with my own observations as illustrated beyond. Variations in the orientation of crystals of tin and of lead occur under certain conditions (Ewing and Rosenhain, "The Crystalline Structure of Metals, *Trans. Royal Soc.*, 1901, A, vol. 195, pp. 279–301; Rosenhain, *Journ. Iron and Steel Inst.*, 1904, No. I, pp. 388 and 335). But then the native crystals of some minerals are often curved, for instance those of calcite, siderite, and pearlspar (curved dolomite).

Beyond this, some have imagined a very narrow region in which the pine-tree tips of adjoining grains interlace, giving rise to a mixed orientation, though, as we have seen, the smoothness of the intergranular fractures tends to show that, at least in many cases, there is no such interlacing.

In that each grain is by definition a mass with uniform orientation throughout, and differing sharply in orientation from its neighbors, it is a structural unit.

346. Meaning of Grain.—It is important to recognize that "grain" is used here in the sense in which it applies to a "grain" of maize or other cereal. It includes the whole of a structural unit, and is not applicable to part of such a unit. Fragments cut from a grain of maize are not grains in this sense, nor are fragments broken from a single uniformly oriented grain of metal. Indeed a grain is a cell, and there are strong arguments for calling it a cell rather than a grain. Nevertheless the word "grain" is so firmly established, and "granular" applies so aptly to an assemblage of such grains, and to a fracture passing along the grain surfaces, that I retain it.

If, as we surmised in the last section, there is an intergranular envelope, that is another point of resemblance between these grains and the organic cells; yet not a necessary one, because some even of the organic cells are thought to lack envelopes.

It is true that "grain" is often applied to objects which are not structural units, such as grains of sand or of salt, each of which may be only a small fragment broken from a much greater mass which was a structural unit. Herein lies an objection to the word "grain" in the specific sense in which it is applied to uniformly oriented grains of metal.

347. The allotriomorphism of the grains, their lack of a true idiomorphic crystalline outer form, is evidently a necessary consequence of the environment (§174). No one crystalline grain is able to develop the geometrical outer shape toward which it tends, because the space which it needs for that development is already occupied by its neighbors on all sides. The crystals of our mineralogical cabinets often owe their perfect outer form to their growing freely in cavities, in which they are deposited for instance from aqueous solutions, or to growing within a relatively soft magma, the mobile particles of which the growing crystal is able to push aside.

This allotriomorphism remains a necessary consequence of the conditions, whether we regard the grain boundaries as self-determining after Quincke, or as determined directly by the interference of the growth from nuclei, and whether we admit or deny the presence of an amorphous cement between adjoining grains.[1]

348. Tripartite Nature of Metals.—Thus the metals are (1) microscopically crystalline, in that each microscopic grain is a true crystal; (2) microscopically cellular, in that they are aggregations of these grains, each of which may be regarded as a cell, probably with an envelope; and (3) macro-

[1] See Desch, "First Report on Solidification," *Journ. Inst. Metals*, 1914, No. 1, vol. 11, p. 69.

scopically amorphous, because the orientation changes from grain to grain, and hence the mass as a whole has no internal orientation of its own. This macroscopic amorphousness is often limited, as when a steel ingot is divided into columnar crystals during solidification, or as when wrought iron or nickel steel is made fibrous or laminated by the drawing out of slag or other impurity in rolling. Again, the scale of crystallization and of the cellular structure, though nearly always small compared with the mass of the metal, varies greatly so that it may cease to be microscopic, as in the cases of the silicon steel shown in Fig. A of Plate 16 and of the crystals shown in the frontispiece.

349. The Crystalline Structure of Iron.—The important crystalline forms for iron are the cube, Fig. 34, the octahedron, Fig. 35, and the {211} or leucite trapezohedron, Fig. 37. Let us here consider them and their relations, to the end that we may be in a position to interpret the directions of the lines and bands which deformation marks out, both on alpha iron or ferrite and on gamma iron or austenite.

The octahedron is important because that is the form in which austenite habitually soldifies, and also because in gamma iron or austenite it is followed by the twinning planes,[1] the slip planes (§443), and the Widmanstättian structure (§366).

The cube is important because it is the form in which alpha iron or ferrite habitually deposits (§342); because the strongly marked cleavage of this ferrite is cubic (§365); because the Neumann bands or narrow twins of ferrite and perhaps also its slip bands occasionally follow cubic planes (§591); and because the mineral rhabdite, the niccoliferous iron phosphide, lies in these cubic cleavages at least in some meteoric iron.[2]

The {211} trapezohedron is important because these Neumann bands of ferrite, and probably also its slip bands, habitually follow the planes of this form (§§445 and 589).

Part of the evidence on which this 211 direction is inferred for the slip bands is based on their directions on rhombo-dodecahedral faces. Hence the rhombic dodecahedron, Fig. 36, assumes for us a certain importance.

Of these planes the cubic ones are the most resistant to etching, the octahedral ones the least resistant, and all other random faces are intermediate, according to Stead.[3]

Though, as the crystals in the frontispiece and in Fig. E of Plate 2 show, austenite, in solidifying, forms octahedra under favorable conditions, and is thus octahedral in nature, and though the ferrite into which such austenite crystals transform in cooling past the transformation range preserves this outward form, it is simultaneously both cubic and trapezohedral in nature.

[1] *Journ. Iron and Steel Inst.*, 1906, vol. 71, No. III, p. 468.

[2] Stead, private communication, Dec. 7, 1914.

[3] Private communications, Dec. 7, 1914, and Jan. 4, 1915. It is true that deep etching in my experiments, while it caused very pronounced figuring on the three cube faces of a specimen of thermit steel, caused little on the three random faces of the same specimen. Nevertheless the quantity removed may have been greater on these random faces than on the cube faces.

That is to say, an octahedral crystal of austenite transforms, in cooling, into a pseudomorphic octahedral crystal of ferrite, and that ferrite has cubic cleavages and 211 trapezohedral twinning planes and very probably 211 slip

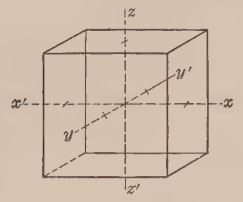


FIG. 34.—Cube {001}.

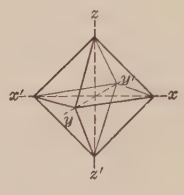


FIG. 35.—Octahedron {111}.

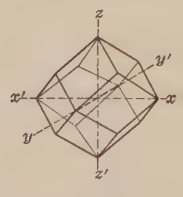


FIG. 36.—Rhombic dodecahedron {011}.

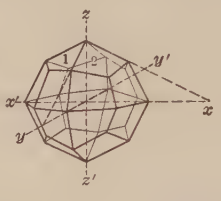


FIG. 37.—Trapezohedron, leucite {211}.

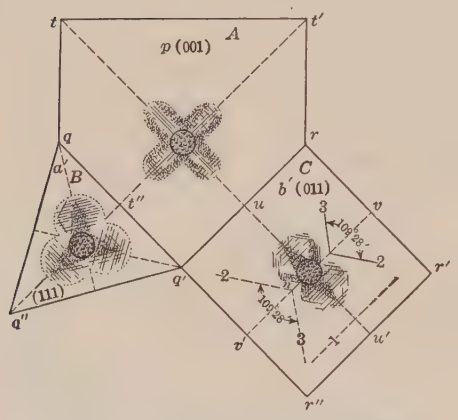


FIG. 38.—Gamma iron or austenite.

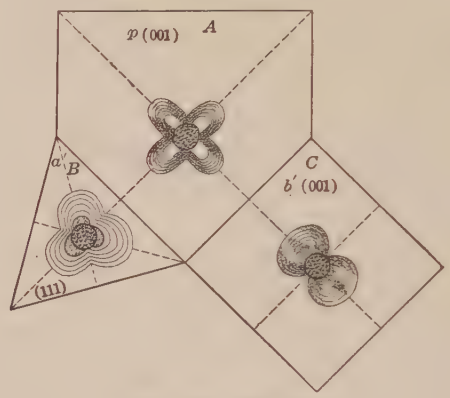


FIG. 39.—Beta iron.

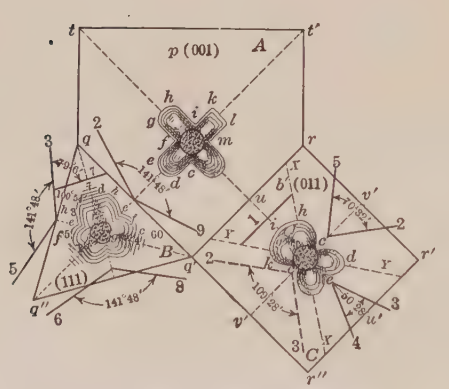


FIG. 40.—Alpha iron or ferrite.

FIGS. 38 to 40.—The deformation figures of iron belong to the isometric system. (Osmond and Cartaud.)[1]

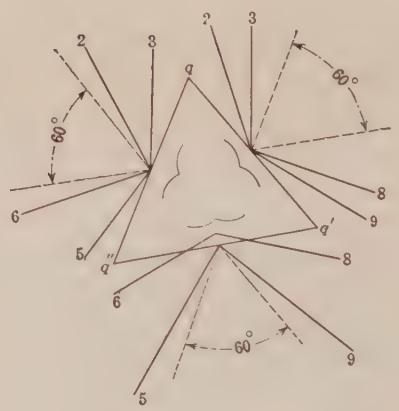


FIG. 41.—Comparison of the straight elements of the silhouettes of face *B* of Fig. 40 with octahedral and 211 intersections of such a face.

planes. After considering the crystalline axes, let us as the next step consider how this ferrite is simultaneously trapezohedral and cubic.

[1] *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, Plate XLIX, opposite p. 448.

350. The Crystalline Axes.—The conception of the existence of these three axes and of their relations to the crystalline forms is of the first importance here, though getting it is a discontinuous process of which the first performance for most of us covers fractions of some days, our subliminal man taking his own gait. It is because we fail to recognize that this gait, like that of the heart and kidneys, is not subject to our will, and because we have

Fig. 42.—Octahedron modified by cube.

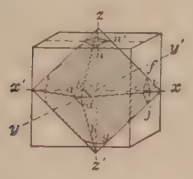


Fig. 43.—Intersections of a cubic truncation of an octahedron, *abcd*, by cubic and octahedral planes.

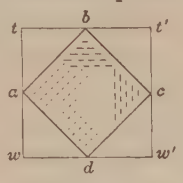
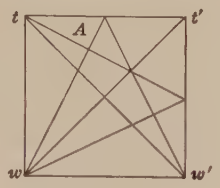


Fig. 44.—Intersections of cube face by 211 planes.



Legend to Fig. 43: Cubic Intersection----- Octahedral Intersection..... Cubic Face = $tt'w'w$,

Fig. 45.—Cubo-octahedron. Cube modified by octahedron.

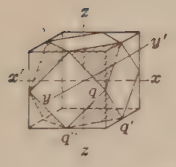


Fig. 46.—Intersections of an octahedral truncation of a cube by cubic and octahedral planes.

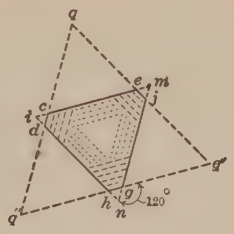
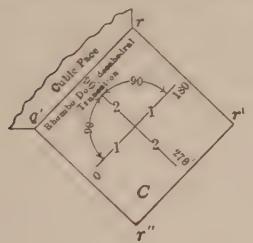
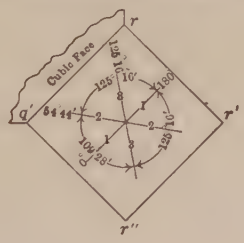


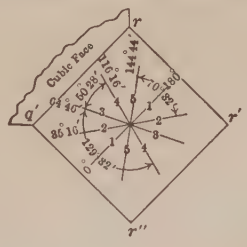
Fig. 47.—Intersections of an octahedral truncation of a cube by 211 planes.

Intersections of a rhombo-dodecahedral truncation of a cube

by cubic planes. Fig. 48. by octahedral planes. Fig. 49. by 211 trapezohedral planes. Fig. 50







not the patience to await his time, that we leave stubborn problems unsolved. Set them to him and he will solve them in his own time. The chief difference between men is in the rapidity and accuracy of their subliminal work, things which for each of us are relatively fixed. Far better a subliminal self who solves aright than one who solves only quickly.

Each crystalline form has three imaginary axes meeting at its center, and, in the case of the isometric or cubic system which alone concerns us here, at

right angles to each other and strictly interchangeable. We may here pass by the hexagonal system in which the number of axes is four. These axes, xx', yy', and zz', and the points at which each intersects the bounding planes of the form, are shown for instance in Fig. 34. Their position is fixed in any given crystal or crystalline grain, but of course varies from grain to grain. Moreover, the relations of the several possible crystalline forms to these axes and through these to each other are fixed.

351. The Octahedron.—If there is an octahedron, or octahedral planes, then they bear to these axes the relation shown in Figs. 35 and 42, each octahedral plane cutting all three of these axes at one and the same distance from their intersection, a distance which for convenience we call unity. The absolute distance of course varies from crystal to crystal and from plane to plane in any one crystal, so that it is called unity only for the purpose of comparing its intercepts on the three axes with each other. But this purpose is of the very first importance, for it fixes rigidly the direction of the plane, and its direction relatively to other planes. The axes thus are imaginary standards absolutely needed for comparing different planes.

Because an octahedral plane has for its fundamental property the ratio 1:1:1 between its intercepts on these axes, it is called a 1:1:1 plane, or briefly 111.

352. The Cube, the Miller Indices.—As shown in Fig. 34, each of the six cubic faces cuts one axis at a distance from the center which is the same for all, and hence is taken as unity. Each face runs parallel to the two other axes and hence cuts them at infinity. Because this is the essential property of cubic faces, they might be called $\infty : \infty : 1$, because that is the ratio between their intercepts on the axes. But for reasons of convenience into which we need not enter, they and the other crystalline planes are represented by the "Miller indices," which are the ratios of the reciprocals of these intercepts. Hence the cube is called 0 : 0 : 1 or briefly {001}. Looking back at the octahedral planes, we see that their name, 111, would be the same whether it represented the intercepts themselves or their reciprocals.

To distinguish an octahedral or a cubic plane from an octahedral or cubic from enclosing space, the plane is indicated by its numerals alone, such as 111 or 001, and the form by those numerals braced together, for instance {111} and {001}, or (111) and (001).

353. Coexistence of Cubic and Octahedral Planes.—Though, as just indicated, the forms which coexist in ferrite are the cubic and the 211 trapezohedron, yet the octahedral form is so much simpler than this latter that I use it for illustrating how this coexistence occurs, on the ground that if the reader sees how these two simpler forms can coexist, he will readily admit that less simple forms also can, the fact of coexistence and not the complexity of the coexisting forms needing explanation.

Conceive that a grain of ferrite, shaped like the part of Fig. 42 which is darkly shaded, is first sliced up into a very great number of very minute, indeed submicroscopic cubes, the faces of all of which are parallel to the cube

faces of this figure, and that these cubelets are next reassembled in their initial positions and cemented together with a cement which is half as strong as the ferrite itself. These cemented planes will then form planes of weakness, planes along which the mass tends to cleave if a chisel, suitably inclined, is forced into it. This assemblage of these cubelets will then look like Fig. 51.

After the mass has been reassembled thus, suppose that it is next sliced up into a very great number of submicroscopic octahedra, by cutting slices along many planes parallel to the faces of the octahedron shown in this figure, and that then all these little octahedra are reassembled in their initial position, and bound together with strong links which yet permit a certain limited degree of sliding to occur. These will constitute octahedral slip planes. Conceive clearly that each grain of metal is thus cut up simultaneously by these seven sets of planes of weakness, three cubic sets and four octahedral ones.

Here each cubelet is supposed to have its own set of axes, but these axes are parallel to those not only of the main octahedron but also of each little octahedron. As before the faces of each little octahedron, like those of the main one, cut its three axes at unity from its own center, while the faces of each cubelet cut one of that cubelet's axes at unity and two others at infinity.

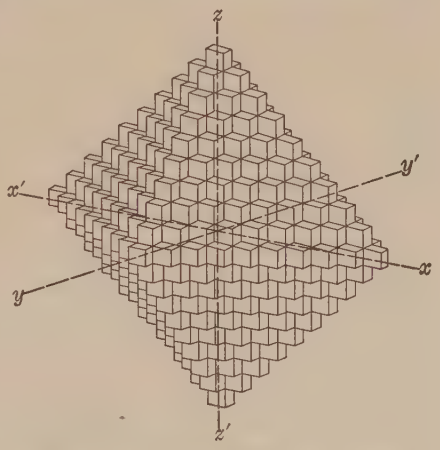


FIG. 51.—Octahedron built up of cubes.[1]

354. This Coexistence in the Polyhedral Grains.—This same relation holds between the coexisting cubic planes and octahedral planes not only in idiomorphic crystals, but also in the allotriomorphic crystals of which our grains of ferrite or of austenite consist. If any such grain, whether of iron or other substance, has simultaneously cubic and octahedral structural planes, whether visible or only capable of development, then it contains three sets each parallel to one of the three pairs of faces of a cube, and four sets each parallel to one of the four sets of faces of an octahedron, and these three sets of cubic planes are inclined toward the four sets of octahedral planes exactly as the cube faces of Fig. 42 are to its octahedral faces. This indeed follows from the specific inclination which each of the two sets of planes bears to their common system of crystalline axes.

If, in addition, the grain contains trapezohedral planes, developed or latent, the relation of these to these same axes and through these axes to the cubic and octahedral planes is rigidly fixed.

Our "space lattice" conception indeed makes it easy to understand that there should be structural planes following various directions, those in

[1] Fig. 2, p. 571, "Encyclopædia Britannica," 1910, vol. 7, p. 571.

each direction having their special characteristics. The characteristics of the cubic planes may differ from those of the octahedral, and both from those of the {211} trapezohedron.

Given the existence of such independent sets of planes of weakness, it will depend on the direction and probably on the rapidity of increase of the stresses, that is, on their suddenness, whether yielding shall occur along one set or another, or simultaneously along two or all three.

355. Reciprocal Truncation of Cubes and Octahedra.—The conceivable division of an octahedron into cubelets, shown in Fig. 51, having served its purpose, it is more convenient henceforth to conceive each cube as composed of a nest of concentric cubes and each octahedron as composed of a like nest of octahedra, all the cubes and all the octahedra being grouped concentrically about one and the same set of axes. We may now fix our attention on a cube, the intercepts of which are slightly less than those of the coexisting octahedron, as in Fig. 42, or very much less, or less in any other degree as long as they are less, so that the cube and octahedron intersect. But whatever this ratio between the intercepts, the relative direction of the octahedral and cubic faces is exactly that here shown. From this it results that each of the six cube faces slices off one of the six apices of the octahedron evenly, and that all six make exactly the same slicing; that each of the eight faces of the octahedron slices off one of the eight corners of the cube perfectly evenly; and that all eight of these slicings too are identical in dimensions.

356. The rhombic dodecahedron or granatohedron, Fig. 36, is formed by the intersection of 12 planes each of which is like an octahedral plane in cutting two of the axes at unity, and differs from it in being parallel to the third axis. Thus if we imagine that the octahedral face ABC of Fig. 122, p. 596, is hinged on BC and swung into a vertical plane, it would become a rhombododecahedral plane, and if now moved parallel to itself from BC to IM it would cut the octahedron along $GHIKLM$ as here shown. Because each plane cuts one axis at infinity and two at unity, its index is $1 \div \infty : 1 \div 1 : 1 \div 1$, or {011}.

357. The {211} or leucite trapezohedron, Fig. 37, is formed of planes each of which is like an octahedral plane in cutting two of the axes at unity, as at BC in Fig. 122, and the third at half of unity, as at A'. Because the three intercepts of this plane are thus in the ratio $1\frac{1}{2} : 1 : 1$,[1] its index is 211. This plane BCA' thus corresponds to the plane 1 in Fig. 37, the intercept of which on the vertical axis zz' is half of those on the horizontal axes xx' and yy', though the foreshortening rather tends to conceal this.

Four like planes thus starting from each end of each of the three axes,

[1] For instance, the inclination of a lamella, ltv, of Fig. 119 which is to cut the top of the cube diagonally at lnv and the front and right-hand face along a line running from their apices l and v to the middle of the opposite edge at t, must be such that the points m'' and m' at which, when produced, it cuts the two horizontal axes xx' and yy', are twice as far from the center s as the point n at which it cuts the vertical axis zz'. Hence the ratio of these intercepts is 2:2:1, and the ratio of their reciprocals is $\frac{1}{2} : \frac{1}{2} : 1 = 1 : 1 : 2$, which is equivalent to the index {211} given above, because the order in which these numbers stand is immaterial.

there will be a total of $4 \times 2 \times 3 = 24$ planes, which will mark off on each other the 24 trapezia by which the crystalline form is bounded, as shown in Fig. 37. This trapezohedron is the crystalline form to the faces of which the Neumann lamellæ are parallel, as set forth in §589 to §591.

358. The Intersections of Crystalline Planes.—The nature and mechanism of the plastic deformation which occurs when iron is strained beyond its elastic limit, that is overstrained, whether in tensile testing, wire drawing, or otherwise, are studied in part by means of the lines and figures which this deformation causes, for instance on a previously polished surface. This study is made to the best advantage when we know what the crystallographic nature of this surface is, for instance when we know that it is parallel to the faces of a cube, and hence is a true cubic plane, or that it is an octahedral plane, or whatever it may be. Moreover, because the directions of the planes of these various crystalline forms relatively to each other are known, the directions of the intersections which the planes of any one form make on the planes of any other known form is known. Hence, knowing that the polished surface of our specimen is for instance a cubic plane, we may be able to infer from the directions of the lines which deformation makes on it whether that deformation is taking place along octahedral planes, or more broadly along what crystallographic planes it is taking place, and indeed whether those planes are or are not crystalline planes. The directions which the intersections of certain crystallographic planes make on the cubic, octahedral, and rhombo-dodecahedral planes are shown in Figs. 43 to 50. Appendix I explains the derivation of certain of these directions for those who may care to verify them.

359. Grains of the First and Second Orders. Ungeometrical Grain Fragments.—Figs. B of Plate 24, B of 34, and G of 36 show what may be called grains of two orders in ferrite, each large grain, or grain of the first order, being made up of several fragments, or grains of the second order, with the usual grain shape and oriented alike. Indeed this common orientation is often the only ready means of distinguishing grains of the first from those of the second order. Thus the whole field of Fig. B of Plate 34 is composed of a single grain, as is shown by the uniform direction of its Neumann bands, yet it is broken up into many fragments or grains of the second order, and so are those of Fig. B of Plate 24. Grain *T* of Fig. G of Plate 36 is broken up in like manner into many fragments, probably by the severe punching which this specimen has undergone, as shown by the inequiaxing of grain *TUVW*.

In some cases these smaller grains may be irregular fragments into which the larger grains have been broken by deformation, quite as a brick wall, in spite of its geometrical internal structure, might well break up into irregular masses. In other cases, as when there has been no deformation which can account for such a breaking up, I interpret these grains provisionally as representing the crystalline growth which, in the transformation of the austenite into ferrite in cooling through the transformation range, has

started at independent centers within any one austenite grain, spreading out from each nucleus till it meets that proceeding from the other nuclei, when they limit each other in the way sketched in Fig. 26, p. 135. It is true that the grains of austenite are smaller than those of the ferrite derived from it in Humfrey's[1] vacuum etching experiments with electrolytic iron at 1,000°, followed by slow cooling. The vacuum heating etches the grain boundaries both of the austenite which existed at 1,000° and of the ferrite[2] into which it transformed in cooling, the latter boundaries being recognizable by their changing the direction of slip bands made after the cooling (§425). But the relative sizes of these two sets of grains may, for all we know, vary greatly and according to laws which may be as complex as those which govern the grain size of ferrite, so that though the ferrite grains are larger than those of their mother austenite under these specific conditions, the reverse may be true under the conditions represented by Figs. B of Plate 24, B of 34 and G of 36.

The fact that all these grains of the second order are oriented alike I refer to their severally inheriting the orientation of the mother austenite whence they spring, or at least bearing a constant relation to that orientation. This interpretation is of course only provisional.

360. The crystalline constitution of carbon steel is complicated by the fact that it differs from the pure metals to which much of what has thus far been said should be confined, in consisting usually of two or more perfectly distinct substances, pure iron or ferrite and iron carbide or cementite, the whole of one of these being interstratified with part of the other as pearlite in slowly cooled metal, as explained in §§70 and 71. Nevertheless this ferrite is composed of grains like those of any pure metal, and so is the cementite.

This is very evident in steel very poor in carbon, for here there is so little pearlite, or so little cementite set free by the divorce of the pearlite, that it does not interfere greatly with the structure of the ferrite. Even in steels with an important carbon content, and hence with an important quantity of pearlite, about and through which the ferrite is grouped as a network and in independent masses, that ferrite can often be seen plainly to be divided thus into polygonal grains, strictly comparable with the grains of ultra low-carbon steel or of a pure metal. The great white network of ferrite in Fig. E of Plate 16, in steel of 0.46 per cent. of carbon, is evidently divided thus into these irregular polyhedra, and so is the ground mass of ferrite in the malleable cast iron of Figs. A, B, and C of Plate 4.

It is true that the individual grains of the little masses of ferrite and of cementite which make up pearlite are so extremely minute that they cannot be detected readily if at all; and that in steels near the eutectoid carbon-content, and hence with but little pro-eutectoid ferrite or cementite, the

[1] *Carnegie Memoirs, Iron and Steel Inst.*, 1912, vol. IV, p. 83.

[2] Rosenhain and Humfrey, *Proc. Royal Soc.*, 1910, A, vol. 83, pp. 200–209; Rosenhain and Ewen, *Journ. Inst. Metals*, 1912, No. 2, vol. 8, p. 149.

masses of this pro-eutectoid element are very often so thin that the individual grains composing them cannot be distinguished.

361. Grains of Austenite and of Pearlite.—Because "grain" thus means a unit oriented uniformly throughout, it can be applied also to masses of austenite and of pearlite.

Above the transformation range the austenite can be shown by hot etching to consist of polyhedral grains exactly comparable with those of ferrite in ultra low-carbon steel, and with those of pure metals in general. Fig. B of Plate 16 shows such austenite grains developed thus, with polygonal boundaries, and with uniform orientation throughout each grain. Hence the term "austenite grain."

In much the same way the dendrites of eutectic and of pearlite follow a constant direction throughout certain regions, as seen in Figs. A and G of Plate 1. While it is true that each of the little dendrites of ferrite in the pearlite is itself composed of separate grains, each with its own orientation, even if these cannot be distinguished, yet these dendrites of ferrite, and those of cementite with which they are interstratified, have uniform direction throughout a given area. Hence each of these areas may reasonably be called a "pearlitic grain," a term which we naturally apply to the polygonal figures in the pearlite of Fig. F of Plate 3. "Colonies" is in some ways better than "grain," as applied to these masses of pearlite the dendrites of which follow a constant direction, in spite of the differences in orientation of the individual grains of which each dendrite is made up (§175).

The case of martensite is less simple. When the austenite, of which the steel consists when above the transformation range, is cooled rapidly by quenching in water, it transforms as far as the martensite stage, in spite of the rapidity of the cooling. Nevertheless the boundaries which, when above the transformation range, had separated the several grains of austenite from each other, are often very prominent in the martensite which results from quenching that austenite, as in Fig. A of Plate 2, and on that account one is tempted to call each of these masses resulting thus from the transformation of one grain of austenite a "martensite grain." We shall see in Chapter 30 that the fracture of such steel is governed by the size of these austenite grains, the outline of which still persists pseudomorphically in the martensite. But within each of these masses bounded by what were the grain boundaries of the austenite, there may be very many different needles and arrow heads of martensite, such as are seen in Figs. A and B, Plate 10, and Fig. A, Plate 11, and "martensite grain" might easily be misunderstood to apply to each of these. Hence the expression should be used cautiously if at all.

362. Network, Kernels or Meshes, and Shells.—In cooling through the transformation range the pro-eutectoid ferrite or cementite expelled by the several austenite grains gathers in the grain boundaries, giving rise to the "*network structure*," consisting of "*kernels*" of austenite and "*shells*" of the pro-eutectoid element. "Kernel" is used here because more suggestive than the term "mesh" used by many writers. The unit composed of a kernel with

its shell may called a "walled cell." This network structure seems to be a special form of the poikilitic structure of the petrographers. Figs. D and E, Plate 1, and D, E, G, H and I of Plate 3 show this structure. The kernels in each case are of pearlite, while the shells are of ferrite in hypo-eutectoid steel and of cementite in hyper-eutectoid steel. These pearlite kernels often contain many small islands, spines, etc., of the pro-eutectoid element.

As I have already explained, the network system is a transient arrangement. Surface tension is always striving to break it down, and in the end to assemble all the cementite in one mass and all the ferrite in another. Nevertheless in many cases the stability of the network is very great (§§235, 236).

363. The banded structure,[1] shown on Plate 43, with longitudinal white bands richer in ferrite and dark ones enriched in pearlite, is related to the network structure. It appears to result from the drawing out of the more phosphoric parts, concentrated between the trunks of the dendrites during solidification, into long bands by the rolling and forging. This will be studied further in §779.

Now that we have taken a bird's eye view of the crystalline forms which concern iron, and of the tripartite nature of steel, we may take up in detail the evidence already outlined of the crystalline nature of the metals in general, and of iron in particular.

[1] Compare Oberhoffer, *Zeitschr. anorg. Chemie*, 1913, vol. 81, p. 156; Rosenhain, *Proc. Internat. Assoc. Testing Materials*, VI Congress, 1912, 1st Section, II, 2, p. 17; Brearley, "The Use of Microscopic Methods," *Sheffield Soc. of Engineers and Metallurgists*, 1909–1910.

CHAPTER 14

FURTHER EVIDENCE OF THE CRYSTALLINE STRUCTURE OF METALS

364. Summary.—This chapter takes up further evidence of the crystalline nature of metals, such as the geometrical shape of the bodies which can be cleft from them, the geometrical disposition of foreign matter within them, the geometrical shape of the etching and pressure figures, the geometrical arrangement of air bubbles, the uniformity of orientation throughout each grain, and the change of orientation from grain to grain. In passing, the motive force of the individual grains which expels foreign matter, the mechanism by which it works, the genesis of the grain junctions, and the reasons why the etching pits retain their geometrical shape during their growth are considered. One of the most cogent evidences of the crystalline structure, the silhouettes or pressure figures composed of the slip bands, will be considered in Chapter 17 after the slip bands themselves have been studied.

365. Geometrically Shaped Bodies Cleft Out.—Figures F and G of Plate 15 show rectangular masses cleft out of thermit iron. Stead[1] has shown even more striking cubes and other rectangular bodies which he cleft readily out of a mass of "very slowly cooled granular phosphoretic iron," containing 0.75 per cent. of phosphorus and no carbon. Here the angles were almost exactly 90°.

366. Metals dispose foreign matter geometrically along their own crystallographic planes, as is shown familiarly by the Widmanstättian figuring common in meteorites, in steel which has been cooled slowly from a high temperature, and in many other alloys.[2] As steel cools through the transformation range, the austenite of which it consists while above that range gives birth to a continuously increasing quantity of the pro-eutectoid element, ferrite in hypo-eutectoid and cementite in hyper-eutectoid steels, and tends to distribute this element in its own cleavages. This distribution of the pro-eutectoid ferrite is shown in Fig. R of Plate 8. A like distribution of the pro-eutectoid cementite is seen in Figs. E, F, I and K of Plate 9.

That this distribution of the pro-eutectoid ferrite is strictly according to the cleavages of an octahedron has been shown conclusively by Belaiew.[3] The distribution of the pro-eutectoid cementite is so closely like that of the ferrite as to leave no doubt that it too follows those same cleavages.

[1] *Journ. Iron and Steel Inst.*, 1898, No. I, vol. 53, p. 179; *The Metallographist*, 1898, vol. 1, pp. 330–331.

[2] Belaiew, *Journ. Inst. Metals*, 1914, No. 2, vol. 12, p. 46.

[3] *Annals of the Imperial Russian Technical Association*, 1909; *Rev. de Metallurgie, Memoires*, 1912, vol. 9, p. 335.

The important point for our present purpose is that this distribution is geometrical and along definite crystallographic planes.

367. Each Austenite Grain an Organism with the Power of Expelling Foreign Matter to Its Surfaces and to Its Cleavages.—We may here consider evidence supporting this theory, which was outlined in §235.

The distribution of the pro-eutectoid cementite in layers parallel to the faces of an octahedron cannot be referred to any motive power in the cementite, because this does not belong to the isometric crystalline system to which the octahedron belongs, but to the orthorhombic system. Hence we refer it to the motive force of the mother austenite itself.

The fact that particles of slag are habitually surrounded by masses of ferrite or cementite points in this same direction. The austenite grains while above the transformation range would, by like process, naturally dispose along their surface or their cleavages any slag which they contain. In later cooling through that range they arrange the pro-eutectoid element then born in like manner, and hence this later arranged matter is found surrounding the earlier arranged slag, which itself is most unlikely to have any such power of arranging its disconnected masses geometrically. Because the crystalline power of the austenite is the only agent which can have arranged the slag thus geometrically, we infer reasonably that it is also the agent which has thus arranged the pro-eutectoid element.

The common occurrence of slag within ferrite masses,[1] shown in Figs. F of Plate 43 and G of Plate 44 is usually supposed to be caused by some attraction which the slag has for the ferrite. But by what piping does this Pied Piper entice the ferrite from within the austenite, draw it across considerable distances, and enwrap itself in it as in a garment? Why create so improbable an explanation when the ejecting power of the austenite would naturally lead to just this stratigraphic occurrence? The ejection of the ferrite would naturally occur only during the cooling through the transformation range, that is, after the expulsion of the slag from the austenite, and this later ejected ferrite would naturally be deposited about the earlier ejected slag.[2] The stratigraphic history of the ferrite bands or ghosts, which are analogous to this present ferrite network, accounts equally well for their often enclosing slag masses (§780).

368. Evidence Supporting this Hypothesis.—From this conception that the occurrence of foreign matter at the boundaries of the grains is due to its expulsion by the grains themselves, it follows that such expulsion ought to cover the surface of any block of steel with foreign matter, if there is foreign matter present capable of expulsion, for the whole of the surface of such a block is composed of the existing outsides of the grains which abut against it, and this I have found to be true in all the cases which I have examined.

[1] Brearley, *Proc. Sheffield Soc. Engineers and Metallurgists*, 1909–1910, p. 56.

[2] Compare the Author, "Life History of Network and Ferrite Grains in Carbon Steel," *Proc. Amer. Soc. Testing Materials*, 1911, vol. 11, pp. 271, 295, 373, and 375; and Remarks in Discussion, *Trans. Amer. Inst. Min. Eng.*, 1913, vol. 47, p. 659.

The first case which came to my attention was that of a steel of 0.40 per cent. of carbon which I sent to the United States Bureau of Standards. After heating it in an air vacuum of usually 0.1 mm. or less of mercury, Dr. Burgess reported that its surface was covered with ferrite.[1]

This at first was taken to mean that there was surface decarburization even in such a high vacuum; but reflection showed me that during the cooling down through the transformation range, the pro-eutectoid ferrite as it formed had been ejected to the outer surface of the grains abutting against the surface, quite as it is habitually expelled to form a network between the grains of austenite through the mass of such a steel. This action I likened to the precipitation of the particles of any crystallizing solid, for instance from an aqueous solution of cupric sulphate, upon all the surfaces of the liquid, including the sides of the vessel, any wooden strips hung down in the liquor to collect the crystals, and indeed the upper surface of the liquor itself, against which fine crystals often form, to be dragged down by gravity as soon as they reach considerable size.

I next showed with A. G. Levy that if the entrance of a hole drilled into a block of steel is closed so as to exclude the air, and the steel is then heated above the transformation range and cooled slowly through it, the walls of the hole become covered with ferrite if the steel is hypo-eutectoid and with cementite if it is hyper-eutectoid.[2] The presence of such ferrite could be referred to surface decarburization, but that of cementite could not be, and hence could be referred only to expulsion by the mother austenite in cooling down through the transformation range.

Stead[3] about this time informed me that, on heating and cooling two pieces of hyper-eutectoid steel of which the ends had been polished and pressed firmly together, he found the line of junction filled by a layer of cementite, thus confirming this theory.

Again, on etching manganese steel with hydrochloric acid at 575° we found its surface covered with a thin film of cementite, which evidently had been expelled thither. On grinding this film away and polishing and etching, the usual cementite needles appeared.[4]

Again, on coating austenitic manganese steel with copper, heating to 600°, which is below the transformation range, in order to enable the pro-eutectoid cementite to coalesce and migrate, and again cooling, we found a layer of cementite between the steel and the copper, as shown in Fig. A of Plate 17. This layer can have come only from the ejection of the pro-eutectoid cementite out of the austenite grains abutting against the surface.

We developed slip bands on the surface of a specimen of austenite manganese steel, which we next heated to 575° in hydrogen and cooled slowly.

[1] Burgess, Crowe, and Rawdon, *Trans. Amer. Inst. Min. Eng.*, 1913, vol. 47, p. 607.

[2] *Idem*, p. 660.

[3] Private communication, Mar. 4, 1913.

[4] These needles have been taken by Sauveur to be martensite. A. G. Levy and I proved that they are cementite, by etching them with sodium picrate, which blackened them (*Trans. Am. Inst. Min. Eng.*, 1914, vol. 50, p. 508; *Trans. Faraday Soc.*, 1915, vol. 10, p. 267).

Most of the slip bands had now disappeared, as is shown in Fig. D of Plate 16, but on polishing and etching, lines corresponding to them appeared, as shown in Fig. F of that plate. The disappearance is referred to the masking of the slip bands by the cementite sheet expelled to the surface on heating. That it was not a direct effacement by the heating is shown by the fact that bands of cementite appeared, corresponding to them in position, on re-polishing and re-etching.

Fig. B of Plate 17 shows how the cementite coats a cavity in a casting of such manganese steel, evidently having been ejected to this position by the grains of austenite abutting against the sides of the cavity. A. G. Levy and I have found such occurrences of slag in long sinuous lines about the idiomorphic crystals of austenite coming from the axial pipe in a large steel ingot, as shown in Fig. C of Plate 17.

Finally, on heating a piece of hyper-eutectoid steel with 1.45 per cent. of carbon to 1,000° in a sealed silica tube, so as to coarsen its austenite, cooling it slowly so that this austenite might have a good opportunity for expelling the pro-eutectoid cementite to the surface, copper coating it, and then polishing and etching a section cut across it, we found its surface covered with a nearly continuous sheet of cementite of the same order of thickness as the cementite network, as shown in Figs. D and F of Plate 17.

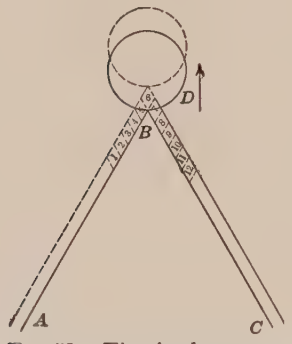


Fig. 52.—Ejection by a growing crystal.

The geometrical arrangement of the air bubbles is shown in §378 to be due to this power of each cell to move the air sidewise along its surface.

To sum this up, the geometrical arrangement of masses of pro-eutectoid ferrite and cementite and of air bubbles is shown to be evidence of the motive power of the austenite grain, and both the automatic coating of austenite surfaces with pro-eutectoid ferrite or cementite, under a wide variety of conditions, and the enclosing of slag masses within ferrite envelopes are shown to be natural consequences of this theory.

369. Mechanism of the Expulsion of the Pro-eutectoid Element.—We can easily form a mental picture of a possible mechanism by means of which this expulsion is brought about, of course understanding that this does not purport to be the true mechanism. Imagine a crystal growing in an aqueous solution to have reached the outline *ABC* of Fig. 52, and now to come into contact with the speck of dirt *D*, the magnification being enormous so that each of the little rhombs 1 to 12 is one crystal unit, the smallest aggregation capable of crystalline structure. As growth continues from the outline *ABC*, new units 1 to 4 and 8 to 12 attach themselves to the old without trouble; but the molecules which are attracted by the region about *B* to form units 5 and 7 find their natural place in part occupied by *D*. The attraction, that form of cohesion which we vaguely call the crystalline force, which causes the molecules thus to seek positions 5 and 7, causes them in

seeking it to wedge D outward in the direction of the arrow, thus giving the liquid access to the very apex of B. 5 and 7 having thus fallen into rank, they in turn attract the particles which are now precipitating out of the solution in contact with their faces to occupy space 6. In crowding into this space they wedge D farther out.

The expulsion of a particle of ferrite from the interior of a crystal of austenite in which it is born, though not quite so simple, may be conceived to be brought about in a generally like way. Assume that a grain of austenite, $LMNO$, Fig. 53, composed of crystal units, the smallest unit capable of the crystalline organization, of which 16 are here numbered, gives birth to a spherical speck of pro-eutectoid ferrite, F. In the spandrels A, B, C, and D about this speck the austenite cannot crystallize in harmony with the surrounding units 1 to 16, because these spandrels are smaller than a crystal unit; or, more broadly, no matter how the crystal units are packed about F, there must be odd-shaped spaces too small to allow such units to form. Thus F prevents the austenite in these spandrels from passing from the amorphous and less stable to the crystalline and more stable state. The attempt of each of these spandrels to crystallize wedges it against F. But spandrels A and B have the advantage over C and D that the resistance to the movement of F upward is less than to its movement in any other direction. Or, looking from another point of view, the mass of units below which is coöperating with units 7 to 12 to force A and B into the crystalline state is greater than the mass of those above which is coöperating with units 1 to 6 to crystallize C and D. Or, to put it more generally, the total crystalline attraction upon A and B is greater than that upon C and D, because A and B are nearer the center of gravity of the crystal. The excess of motive force acting on A and B over that acting on C and D gives rise to a resultant force tending to expel F, even though this means moving units 2 and 3 apart.

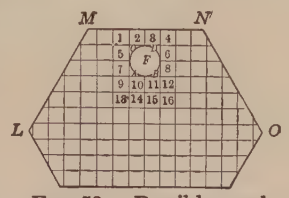


FIG. 53.—Possible mechanism of the expulsion of foreign matter from a crystalline grain.

That growing crystals actually do exert measurable mechanical pressure has been shown. Thus a crystal of alum growing upward in a flat-bottomed vessel has been found to exert an upward thrust of 1 kg.[1]

370. High Organization of a Crystalline Grain.—We are told that organic matter differs from inorganic in its ability to select and assimilate appropriate matter and reject that which was inappropriate. But a crystal growing in an aqueous solution evidently has a power which fits these words, because it, too, selects from the aqueous solutions the matter like itself, and rejects that unlike itself.

The austenite grains display this power of rejection in a more striking way. The selection and rejection by a crystal growing in an aqueous solu-

[1] Desch, *Journ. Inst. Metals*, 1914, No. 1, vol. 11, p. 104; J. W. Cobb, *Journ. Soc. Chem. Industry*, 1907, vol. 26, p. 390. The cracking of cylindrical glass crystallizing dishes by certain kinds of crystals may be connected with this thrust.

tion accompany the birth of the crystalline state, in the sense that the selection occurs only at the growing face of the crystal. But the austenite grain, in ejecting the pro-eutectoid element, or the slag, to its outside, accomplishes this ejection at a later and often at a much later period, after it has itself been in uninterrupted existence for hours or even days. There is no reason to doubt that it would continue to have this ejecting power years or even centuries after its own formation, as in the case of meteorites. This function of the austenite grain in moving this foreign matter considerable distances across its own body to its own surface, or in distributing it in its own cleavages, is a higher manifestation of the crystalline organization than the mere rejection of foreign matter at the face of a growing crystal in an aqueous solution, because the latter may be regarded as simply an incidental and inevitable consequence of the cohesion of the newly depositing crystal units to the existing face of the crystal. Here the newly deposited layers are accepted by the surface layers only; but the ejection of foreign matter by austenite implies the coöperation of many layers throughout the already solid mass, and is comparable with the vomiting of unwelcome matter by a living organism, an act which suggests rather a high degree of organization.

It is quite true that, after this ejection has progressed to a considerable degree, so that the pro-eutectoid ferrite, for instance, has already been assembled into masses of considerable size by the austenite, these coalesced ferrite masses are likely to be connected outside the plane of the microsection by ferrite threads. At this stage the surface tension of these coalesced ferrite masses tends to draw them together, and may well aid the austenite grain in its task of ejection and cleavage massing. But in the beginning the ferrite comes into existence as scattered and disconnected molecules. Till the stage is reached when the movement of these scattered masses has established connecting threads, it is the mother austenite that must supply the motive force.

371. The geometrical etching figures which form in etching some metals are direct proof of crystalline structure. Very striking illustrations are Stead's cubic etching figures of siliciferous ferrite, Fig. G of Plate 17, and Wark's etching figures in austenite etched hot, Fig. B of Plate 16. Stead likens this structure to that of the slates on a roof. After we have seen these we recognize the more easily that the etching figures in common ferrite are also geometrical, though less clearly marked. Such figures in ferrite are seen in Figs. B of Plate 4, T of Plate 9, E of Plate 16, and B to E of Plate 18. They contrast strongly with the smoothness of cementite after etching, as seen in Figs. F and G of Plate 1, L, M, N and O of Plate 3, A to F of Plate 11, and A, B, C of Plate 12. Indeed, this smoothness of cementite and roughness of ferrite often serve to distinguish them from each other.

Other clearly geometrical figures are the grooves etched on three faces of a rectangular block of thermit steel, Figs. E, F, and G of Plate 35. The edges of the block are indicated by broken lines, because they were too far out of focus to be shown photographically. Here we have proof of geo-

metrical structure, in the geometrical relation of these etching grooves to the rectangular cleavage faces of this block. The grooves make with the sides of these faces approximately the angles of 90°, 45°, 26° 34′, and 63° 26′, which last three, as we shall see in §589, are the intersections which the planes of the {211} trapezohedron make with the sides of the faces of a cube. Other geometrical etching grooves of Neumann bands are shown in Figs. A and F of Plate 34. It will be shown in §590 that the grooves of Fig. A follow approximately all of the eight possible directions of the Neumann bands.

The deeply etched surface of a piece of coarse-grained silicon steel is seen, even by the naked eye, to be made up of a great number of very small inclined facets, such as are sketched from *A* to *C* in Fig. 54. Let *ABGC* represent a section cut normally to the face of such a sheet; *AB* the upper surface of one grain; and *BC* that of an adjoining grain. Because the cubelets of *BC* are inclined to the plane of the sheet, etching converts its initially smooth surface into a series of sawteeth. Because those of *AB* have their upper surface parallel to *AB* the etching, in eating away the initially smooth surface *AB*, cuts it into a series of cubic pockets and studs. Stead found that these pockets are extremely small when the etching begins, but later are eaten out into larger ones,[1] which always remain cubic. Such rectangular pockets are shown in black in Fig. H of Plate 17. Every student must notice that the marble, by means of which he generates carbonic acid in his laboratory experiments, is in like manner eaten away geometrically.

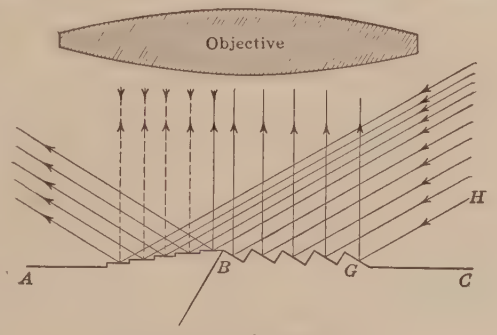


FIG. 54.—Inclined facets.

This observation of Stead's concerning the cubic stud and pits in ferrite means that there are geometrically arranged surfaces or masses within the metal which etch more slowly than the rest, or in short are specially resistant. Such resistant masses are shown clearly in Fig. E of Plate 17, suggesting two sets of outcropping dykes, cutting each other at an angle of about 60° in the plane of this section, but their inclination to each other in space is probably either cubic or that of the {211} trapezohedron. Their projecting above the etched surface, like the projection of harder dykes above the surface of the softer rocks which they cross, means that they are more resistant than the surrounding metal.

The geometrical arrangement of etching pits in ferrite is shown clearly in Fig. D of Plate 18. They are similar and similarly oriented in any one grain of ferrite, but their direction changes from grain to grain, so that if a pit crosses a grain boundary the direction of its sides changes at this intersection. Thus in the upper large grain *A* they are rectangles, in the middle grain *B* they are nearly equilateral triangles, and in the lowest

[1] *Journ. Iron and Steel Inst.*, 1898, No. I, vol. 53, p. 175.

grain *C* they are roughly similar rounded rectangles. The pit on the boundary between *A* and *B* conforms with the pits in both these grains, its left-hand side being a fragment of a rough rectangle and its right-hand side a fragment of a triangle, each fragment being oriented like the pits in the grain in which it lies. Fig. 54A has a like teaching.

The sections of these pits in ferrite are those of cubes and octahedra.

Cogent evidence of the geometrical arrangement of the etching pits is given by Humfrey,[1] who, in studying a very large and regular dendrite of lead, found that the cubic sides of the cubo-octahedral etching pits were parallel to the main axes and branches of the dendrite itself. As these cubic sides are by definition parallel to the crystalline axes of the etching pits, this means that their axes were parallel to those of the dendrite in which they occurred.

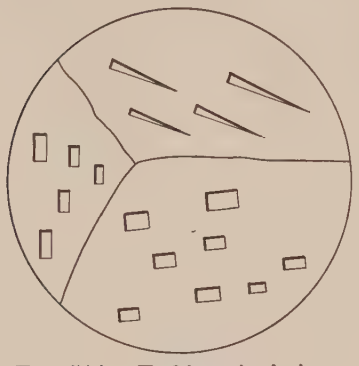


FIG. 54A.—Etching-pits in iron. Desch.[2]

Again, Stead[3] found that a round hole, punched in a cubic face of ferrite, when deeply etched became square and parallel to the adjoining square figures developed simultaneously, as shown in Fig. G of Plate 18. This supports his inference that the cubic faces are the most resistant to corrosion.

372. The Etching Figures are Most Regular on Crystallographic Planes.—It is probably true of iron as it is of native minerals that the sharpest etching figures are made on crystal faces and definite crystallographic planes,[4] and that only moderate sharpness is to be expected when they are made on random surfaces even if these are highly polished. This agrees with the fact that well-marked etching figures in ferrite are usually either squares or equilateral triangles, both of which are traces of the cubic cleavage, the former on cube faces and the latter on octahedral faces. Hence one interprets the fact that well-marked etching figures are rarely other sections of the cube as an application of this general law, the etching not being sharp on the random sections on which the cubic cleavages would make such other figures.

[1] *Phil. Trans. Royal Soc.*, 1904, A, vol. 200, p. 225.

[2] "Metallography," Longmans, Green, and Co., 1910, p. 189, Fig. 71.

[3] Private communication, Dec. 5, 1914. This needs further study. In Fig. A of Plate 34, which shows approximately a cubic face, as is indicated by the directions of the Neumann bands, the crude cross developed by pressure runs approximately diagonally, as it should (Fig. 40). But the round hole formed by the pressure at the junction of the cross arms, on etching tended to develop straight sides parallel rather to the diagonals than to the sides of the cube face, that is to the octahedral rather than to the cubic intersections. The shape of Stead's etching figures in Fig. G of Plate 17 is evidently cubic and not octahedral, because each of the projecting corners is evidently the intersection of three planes, as in the cube, and not four, as at all octahedral corners. Yet it is possible that the etching may sometimes follow the octahedral instead of the cubic planes. The rectangularity of the pits on a cube face is equivocal, because the octahedral interesections of such a face also are rectangular (§836).

[4] Rosenbusch, "Microscopic Physiography of Rock-making Minerals," Iddings' translation, Wiley, New York, 1880, p. 97.

373. The Regularity of the Etching Pits after Great Plastic Deformation.—The etching pits in greatly deformed regions, though they are confused according to Heyn,[1] whose special experience gives his words great weight, yet sometimes show a surprising regularity.

Thus those in steel of 0.40 per cent. of carbon, shown at *A*, *B* and *C* of Fig. B of Plate 18, are regular in spite of the extreme deformation by punching, which drew the ferrite in which these pits occur into the long thin hairs shown in Fig. A of Plate 18. It is true that these rhombs are too long and narrow to be the sections of cubes; but Fig. G of Plate 17 shows that the etching even of undeformed ferrite may well yield rectangles rather than cubes.

Another striking case is shown in Fig. E of Plate 18, in which square pits have been developed by etching ferrite after it had been rolled cold from an initial diameter of ¾ in. to one of ½ in. This would lengthen the grain, which occupies nearly the whole of the micrograph, by about 125 per cent., and reduce its sectional area by about 56 per cent.

Further evidence that the orientation has remained uniform is that, both in a longitudinal and in a cross-section of this bar, the brightness was uniform throughout any given grain but varied from grain to grain (see §380). A longitudinal section showed that the drawing out of the grains naturally expected actually occurred.

The etching pits are square and regularly oriented in certain of the grains of ferrite in Chappell's[2] hoop iron which has been reduced in area 60 per cent. by cold rolling. This is true of all or nearly all the grains in this material which, on reheating to 570° or 600°, failed to recrystallize into minute grain fragments, of course retaining continuity.

374. What do the Etching Figures Represent?—When a soft or porous rock dyke is eaten away by sea or frost, leaving the harder or more compact stratum which it crosses standing out boldly, we find a ready explanation in the difference in hardness or compactness. There must be an equally potent cause of the etching pits which form in metals and other crystalline bodies. We naturally postulate some slight difference of potential. The geometrical shape of the etching figures implies that this difference itself is geometrically arranged, and represents some slight difference in the deposition of the crystalline masses from molten or from solid solution.

The sharp well-marked dykes of Fig. E of Plate 17 suggest that they are composed of metal which is more resistant than the surrounding metal to the etching fluid; and the convexity of the corners of the crystalline blocks of Fig. G of Plate 17 suggests, in this connection, that the excess of resistance inheres not in the surfaces of these blocks alone but in the whole of the material of which they are composed, because otherwise once the surface resistance was eaten through, the interior ought to be eaten down concavely.

[1] "Ueberblick über den gegenwärtigen Stand der Metallographie," *Zeits. Verein Deutsch. Ing.*, 1900, vol. 44, p. 142, col. 2, and Fig. 5, Plate 4.

[2] *Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, p. 467.

But the rectangular and often cubic shape of the etching pits, for instance in Fig. G of Plate 18, tends to sharpen our conceptions. Any given set of approximately cubic etching pits in a given specimen represents only the contour which corrosion has caused up to the instant when it was arrested by our washing off the etching liquor. Had the arrest come earlier or later, a like system of cubic pits would have been found, smaller in the former case, larger in the latter. Hence, to confine our attention to the pits for the moment and leaving the studs out of account, the corrosion proceeds not simply by eating out pits which finally become cubic, but in eating them out in such a way as to keep them at all times cubic. For instance, if Fig. 55 is a vertical section through a mass of metal which, by etching on its upper surface *ADEH*, has been eaten into pits *ABCD* and *EFGH*, the fact that throughout the course of the etching each pit remains at all times approximately cubic, with relatively sharp salient angles, *A*, *D*, *E*, and *H*, and apparently very sharp re-entering ones, *B*, *C*, *F* and *G*, seems to imply that, before the etching began, the space which is now a pit was filled by material arranged as a nest of concentric hollow cubes so constituted that the eating away of any fraction of any individual hollow cube precipitated the immediate and rapid removal of the rest of it, a removal so complete that at every instant while it is going on the interior of the pit is approximately cubic, with straight sides and bottom.

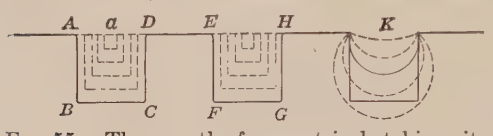


FIG. 55.—The growth of geometrical etching pits.

Thus we come to set aside our natural inference that the reason why the corners *A*, *D*, *E*, and *H* are salient is that they have special resistance to corrosion, because were that all, then during the growth of the pit its walls would pass through a series of shapes such as are sketched at *K*, and there would be no reason why at any given instant they should be straight rather than scooped out in any one of these curved shapes.

375. Does the Retention of the Cubic Form during the Growth of Etching Pits Result from a Common Initial Concentric Cubic Grouping?—We are tempted to suppose that the central cubelet of such a pit, such as *a* of Fig. 55, is for some reason specially attackable, and that it is nested about, like the central beaker of a nest, by concentric hollow cubes so constituted that each stands or falls as a whole, as if each beaker were resistant to attack from inside but not to attack on its outside. The cracking of the inner face at any point gives the etching liquid access to the unresisting back, and the beaker quickly falls as a whole, while its next outer neighbor stays intact till its inner face in turn gives way.

Our temptation is increased by the existence of two differences between adjoining hollow concentric cubelets. The first is that the act of solidification deposits the iron in concentric layers, each a little richer in carbon, phosphorus, and sulphur than the preceding. The second is that the crystallization of each grain proceeds from a single central nucleus, round which the successive concentric crystalline layers may be conceived to group

themselves like a nest of beakers which happen to be cubic instead of cylindrical, each layer bearing, as a layer, a special relation to its neighbors.

These considerations would apply equally to any other form of etching pit. The cubic form has been selected here only to fix our ideas (see footnote[3] at end of §371).

A difficulty might arise from the belief that the center about which etching is concentric is not prefixed but determined by our volition, so that wherever we happen to start a pit, there is the center of the etching growth. But the growth of the pit, while retaining its rectangular or even cubic shape, is probably not concentric about a point thus arbitrarily selected, for it seems to be faster on certain sides of the pit than on others, and thus is probably centered about a prefixed point.

376. Directness of the Proof of Crystalline Structure from the Geometry of the Etching Figures and Pits.—Because it cannot be anything in the nature of the etching reagent[1] which causes the shape of the etching pits to remain constant throughout each grain of the metal and to change from grain to grain, it must be the geometrical structure of the metal itself.

Such etching figures and pits form on the surfaces of non-metallic crystals, and indeed often testify more unequivocally than the external form of a given crystal as to its true nature. The planes which bound these figures are governed by the crystalline laws which determine the outward shape of the crystal, and hence may be used to determine the crystalline structure of masses which lack definite outer shape, not only in the case of non-metallic minerals but also in our present case of metallic masses.[2]

377. Geometrical Arrangement of Air Bubbles.—Smooth surfaces of the more fusible metals, such as cadmium, obtained by casting them on plane surfaces such as those of glass or mica, may be made to disclose the crystalline structure of the metal without etching. In casting thus, the air which adheres to the surface of the glass is trapped, and is then expanded by the heat from the metal so as to occupy considerable space. The various individual grains of the metal as they grow, each from its own nucleus, displace this air like any other foreign body, expelling some of it as far as the grain boundaries, where it collects to widen these boundaries out into rather broad channels, and catching some of it within the grains at interior symmetrically arranged spots, where it forms gas bubbles. In any given grain these bubbles are nearly similar and symmetrically arranged, but their shape and arrangement, like those of the etching pits, vary from grain to grain.

[1] It is true that certain reagents, for instance Heyn's and nitric acid, tend more strongly than others, such as picric acid, to develop regular etching figures. This, returning to our beaker simile used in §375 is to be interpreted as meaning that the difference as regards resistance to etching between front and back of a given layer is more marked as regards the former than as regards the latter class of reagents. The pre-existing structure is developed more truthfully by one than by the other; it cannot be originated by either.

[2] Baumhauer, "Resultate der Aetzmethode in der krystallographischen Forschung," Engelmann, Leipzig, 1894. V. Goldschmidt, *Bulletin*, University Wisconsin, 1904, vol. 3, p. 23. Bayley, "Elementary Crystallography," McGraw-Hill, New York, 1910, p. 220.

Because this geometrical arrangement of the air bubbles, and the change in their arrangement from grain to grain, cannot possibly have been caused either by the air or by the glass on which the metal is cast, they must have been caused by the metal itself. Hence their geometrical arrangement and shape result from the geometrical internal structure of the metal, a structure which is the essence of crystalhood.

It is because this internal structure is geometrical that it thus arranges these air bubbles geometrically and gives them geometrical shape. The natural shape of an air bubble is spherical. The pressure of the molten metal upon it because of gravity would only flatten it down, leaving it circular in horizontal section. The departure from this circular section to the hexagonal section of the bubbles in cadmium is that of a fluid enclosed in a hexagonal mould. Here the metal must be regarded as a mould in which the air is caught. As long as the metal is molten the mould like the air tends toward the spherical shape, flattened by gravity into an oblate spheroid. As the metal crystallizes, the geometrical structure of this crystalline state asserts itself. The surface-tension effort of both metal and air to retain the spherical shape is here outweighed greatly by that crystalline result of the cohesive force of the metal which gives it geometrical structure, indeed outweighed so greatly that it results only in rounding off the corners of the hexagons.

The victory of crystallization over surface tension is indeed usually incomplete and is replaced by a compromise, the air bubbles being neither completely geometrical nor truly circular.[1]

378. The Air Bubbles are Negative Crystals.—These air bubbles must be regarded as internal or negative crystals. On reflection we see that the very geometrical arrangement of the metal which sometimes manifests itself by creating idiomorphic crystals, and thus results in geometrical external shape, if enabled by the presence of an internal cavity to manifest itself internally, naturally results in creating an internal or negative crystal, quite as the assembling of the rafters which give the roof its outward shape gives its interior an exactly corresponding shape, if plasterings and other internal coverings are stripped off so as to disclose it, or as the geometrical structure of brickwork is shown clearly by the shape of the cavity left on pulling out a whole brick.

This present fact that the self-assembling of the solidifying and crystallizing particles of metal leads to the ejection of the entrapped air toward the grain boundaries or to certain geometrically arranged planes, somewhat as the self-marshalling of soldiers into companies in column would tend in part to eject civilians completely from the column, and in part to concentrate them in the spaces between companies, shows that it is the enclosing crystallizing metal which is the motive agent for expelling extraneous matter. The same force which leads such troops to eject civilians standing in the place where a column is assembling would also lead them to eject any intruder

[1] Ewing and Rosenhain, *Phil. Trans. Royal Soc.*, 1900, A, vol. 193, p. 359.

lowered from an airship and so wedged into ranks which were already assembled. Here is additional reason for regarding the mother austenite as the motive agent which ejects to its own grain boundaries or concentrates into its own crystalline planes the pro-eutectoid ferrite or cementite generated within it on cooling into the transformation range.

379. Geometrical Shape of Pressure Pits.—Fig. F of Plate 18 shows how round holes, punched on a cube face of iron, close down to squares when the punched face is next pressed on a smooth surface. This evidence, given by Stead, of geometrical internal structure is as striking as unexpected.

380. That the Brilliancy is Uniform throughout Each Grain but Varies from Grain to Grain Indicates Geometrical Internal Structure.—In Fig. A of Plate 22 the brightest grains are *A*, *B*, and *C*, the darkest ones *D* and *E*.[1] In Fig. B, which shows the same field with the light differently inclined, this order is exactly reversed. Moreover, in both fields the brilliancy is uniform throughout each grain. The reason for this reversal of brightness is evident. If an etched microsection such as that of Fig. 54, p. 285, is examined under vertical illumination, grain *AB* looks bright because its various facets are all horizontal, and hence all reflect back into the microscope the vertical light which they receive from it. But *BC* looks dark because its facets are inclined, and hence cannot reflect any of this vertical light back into the microscope. If the vertical light is replaced by light from the direction *HG*, this will be reflected vertically into the microscope from each of the little inclined facets of *BC* but not from the horizontal facets of *CD*, and hence *BC* will look bright and *AB* dark, reminding you that the mirror with which the mischievous urchin reflects sunlight into your eyes was exactly as bright before he brought it to the annoying angle as at the moment when it blinds you; that the windows golden at sunset are really no more golden then than at noon; and to your humiliation that your own moments of sacrifice and high resolve may only reflect external influences.

The exact uniformity of the brightness throughout each grain of unstrained metal shows that all the facets of that grain must be inclined at exactly the same angle to the light, and hence that they must be exactly parallel, a parallelism which could hardly exist without true internal geometrical structure. This parallelism within a given grain is, of course, a result of the like etching of the various parts of the grain. Such differences in brightness between adjoining grains do not exist in polished but unetched sections.

It is for like reasons that certain spots in a coarse crystalline fracture are bright. It is rare that such bright spots are in one single plane; in general they consist of a great number of parallel or nearly parallel planes all of which are thus favorably disposed for reflecting light to the eye. It is in the fracture of overheated metal that we see such large bright areas, because the overheating has formed coarse austenite grains of uniform orien-

[1] That adjoining grains of gold thus differed in brightness was pointed out by Arnold and Jefferson in 1896; *Engineering*, 1896, vol. 61, p. 140.

tation, each of which in slow cooling has yielded either a single correspondingly large ferrite grain, oriented uniformly throughout, or else a group of nearly similarly oriented ferrite grains. Either condition might cause rupture to follow large plane or nearly plane surfaces, and thus to expose such large surfaces in the fracture, or in short might cause a coarse fracture. So with the coarse cementite plates in spiegeleisen. Each mirror face need not be a single plane, but a large area throughout which all the planes are parallel.

CHAPTER 15

GENERAL CONSIDERATIONS ON DEFORMATION

381. Deformation and rupture may well be surveyed in a general way at this point, before we take up the evidence which their paths furnish that the metals are crystalline.

382. Intergranular and Trans-crystalline Deformation.—We have seen in §345 that metals are made up of grains each of which is a crystal in all save outward form. Deformation and rupture may occur along the boundaries between adjoining grains, in which case they are *intergranular*, or across the body of these grains, avoiding their boundaries, in which case they are *trans-crystalline*. Intergranular rupture yields what is commonly called a "granular" fracture, while the term "crystalline" is applied to certain types of fracture which pass across instead of between the grains, though not all such fractures are called crystalline. In discussing deformation I use the corresponding and unambiguous expressions, "intergranular" and "trans-crystalline." Fractures in fact are always trans-crystalline in properly treated steel, an intergranular fracture being clearly pathological, as for instance in burnt steel.

383. Elastic and Plastic Deformation.—Stress within the elastic limit causes *elastic deformation*, from which the metal recovers its size and shape exactly, after the release of the stress. Stress beyond the elastic limit continues to increase the elastic deformation, and apparently at the same rate as before, but adds *plastic deformation* to it. On the release of the stress the metal recovers from so much of the existing deformation as is elastic, but retains that which is plastic. Thus we recognize the plastic deformation as *permanent set*. "Plastic deformation" in the cold is identical with "cold work" (French *écrouissage*, German *Kaltrecken*) and "overstrain." Stress pushed far enough beyond the elastic limit causes rupture. The fact that every additional refining of our methods of measurement lowers the observed elastic limit ought not to suggest that there is none, any more than inability to detect with exactness the junction of any curve and tangent should suggest that the tangent too is curved.

384. Fluid and Block Movement.—The swirl from my oar seems to start every molecule of the neighboring water into movement relatively to every other molecule.

> "And the startled little waves that leap,
> In fiery ringlets from their sleep."

So with steam expanding in the cylinder, with glass on the pontil, clay on the potter's wheel, and putty and dough when kneaded, ubiquitous stress causing ubiquitous and almost necessarily irregular shear.

Sharply distinguished from these "*fluid movements*" are what might be

called "*block*" movements, that is the movements of whole blocks, the parts of each of which retain their relative position during the motion, as the parts of the earth retain theirs during its rotation. Such motion is like that of a book pulled out from a full shelf, a card from a pack, or a brick from an imaginary wall laid with asphalt.

385. Crystalline and Noncrystalline Movements.—Block movements may in turn be crystalline or noncrystalline. For though the crystalline structure of metals may well lead to strictly crystalline movements, that is to movements along definite crystallographic planes, yet it is compatible with movement along random surfaces, as the rupture of a masonry mass may avoid its joints.

For instance, when a mass of low-carbon steel is deformed plastically, certain of its grains may slide past each other so that their boundaries form a pattern on its surface as in Figs. E and G of Plate 21, in which case the movement is *intergranular*. Or, as when half a pack of cards slips past the other half along the bounding faces of two adjoining cards, the faces of each individual block which moves past its neighbors within any one grain may consist of definite parallel crystalline planes, as suggested in Figs. 63 and 64, p. 336, in which case the motion is by *slip*. Or the particles which compose certain of the blocks, while retaining their relative distances from each other, may all rotate through a definite angle, as when the slats of a Venetian blind are turned, as suggested in Figs. 68, 94 and 104, in which case the movement is by *rotation*. If this rotation is through such an angle that the new position is symmetrical with the old, this rotation is called *twinning*. Cases of such rotation in metals are very common, and are referred to twinning, though strict proof of this symmetry has not yet been given.

Yet in spite of the crystalline organization which slip and twinning imply, deformation and rupture might avoid these crystalline planes, and be wholly irregular, or if regular they might have the regularity only of the lines of surf on a flat beach or of a mackerel sky. They need not be either strictly straight or strictly parallel, and they need not correspond closely if at all to any definite crystalline planes. In this case the block movement is *noncrystalline*.

Not only may these four types of block movement be superposed, but they may be accompanied by fluid movement. That is to say, even if the major part of each of two slices moving past each other remains a crystalline block, and moves like the muntins and panes of one sash past another, yet the metal along their sliding contacts may become decrystallized and revert to the amorphous state.

386. Recapitulation of Movements.—These types of motion may be recapitulated thus:

Types of Motion

- Fluid
- Block
  - Noncrystalline
    - Intergranular
    - Movement of irregular grain fragments
  - Crystalline
    - Slip
    - Twinning
      - Annealing (broad twins)
      - Mechanical (Neumann lamellæ or narrow twins)

These block movements, whether superposed or single, because they occur in a sense independently in the various crystalline grains, may integrate into what, if viewed on a large scale, is equivalent to fluid movement, quite as in the movement of a true fluid we may conceive that the atoms in a given molecule do not change their relative position, so that here movement which is of the block type as regards the atoms within the molecule, that is on the atomic scale, integrates into movement which is fluid when seen on a molecular scale. Every molecule moves relatively to all its neighbor, but the atoms move relatively only to their neighbors in other molecules, retaining their relation to the other atoms in their own molecule unchanged, or at least unaffected by motion which the fluid as a whole undergoes.

387. Nomenclature. Rotary and Vectorial Movement.—The movements of the pirouetting skater are rotary, those of the compass needle in his pocket are vectorial,[1] if we conceive that the needle is only quick enough. He changes not only his position in space but the direction in which he faces, whereas though the needle changes its position in space it retains its direction.

Fluid movements are habitually rotary, as in the swirl of water. Intergranular movements too might be rotary, in the sense that one grain as a whole might rotate relatively to one or more of its neighbors. The trans-crystalline movement by twinning is rotary through a fixed angle.[2] Thus fluid motion and twinning are both unvectorial, with the difference that the unvectorialness of twinning is geometrical and discontinuous, and that of fluid motion ungeometrical and continuous. This distinction between continuous and discontinuous rotation or unvectorialness is important.

388. Slip.[3]—Here a difficulty in nomenclature arises. The name "slip bands" is firmly attached to certain lines which deformation develops on a previously polished surface. Here "slip" naturally implies translation without rotation, so that the very name "slip bands" itself begs the question as to the nature of the thing named, a question to which we are now going to seek

[1] A vector is a line of fixed direction and length, but not of fixed location. Hence "vectorial" may be used to denote this fixity of direction, persisting through movements whether transverse or longitudinal, as for instance the north-pointing of the compass needle.

[2] Of course if deformation goes so far that any given grain is broken in two, its fragments are likely to behave as if they had never been connected, and therefore to change orientation relatively to each other.

[3] Mügge showed in 1895 that this translation, as distinguished from twinning, occurs in non-metallic minerals, and in 1899 that it occurs in metals. He called it "translation," and the markings which such translation causes on a previously polished surface "translation banding" (*Translationsstreifung*), a name still used in German. His experiments left some doubt whether slip, that is actual translation, occurs in iron. An interesting outline is given by Heyn in Martens-Heyn "Materialenkunde," II, A, Springer, Berlin, 1912, p. 219, and *Zeits. Vereins deutscher Ingenieure*, 1900, vol. 44, pp. 433 and 503. This unrotating method of movement was re-enunciated at about the same time and independently by Ewing and Rosenhain, who called these translation stripes "slip bands," the name by which they are now always called in English ("Experiments in Micrometallurgy," *Proc. Royal Soc.*, 1899, vol. 65, p. 85, and "The Crystalline Structure of Metals," *Trans. Royal Soc.*, 1900, A, vol. 193, p. 353). The proposal of Osmond and Cartaud to call them "lines of translation" (*Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, p. 447) has not been followed in English, though in French they are often called by a corresponding name "lignes de glissement" (Robin, "Traité de Metallographie," Hermann, Paris, 1912, p. 153).

an answer. If we use this name "slip bands" and thus acquiesce in this begging, we thereby embarrass the discussion, especially if we call this vectorial translation "slip." Osmond tried to avoid the entanglement by calling these lines "lines of translation." Were this term faultless it might well be adopted. But because "translation" might well be held to include naturally all the six forms of movement except twinning, and because even twinning in the case of the Neumann bands may cause translation which in turn causes definite lines on a previously polished surface, to restrict "translation" thus to this one specific form of translation seems wholly arbitrary, and as likely to cause confusion as the use of "slip bands" is.

These so-called slip bands certainly seem to represent a definite entity, or an as yet unresolved group of entities, which deserves a distinctive name, though following the mineralogical plan this name should make no implications before they are justified firmly. "Slip bands" is improper till slip has been proved.

No really good escape from the embarrassment suggests itself. The best that I can think of is to acquiesce in the use of "slip bands" on the ground that if not the *de jure* it is the *de facto* name so universally used and so likely to be justified by later investigation that to try to disturb it now would be inconvenient and unwise, yet without admitting beforehand that slip bands really represent slip.

In short, I speak of "slip bands" under protest, while asking whether they really are caused by slip, that is by vectorial translation.

Slip, when combined with the rotation of a given grain as a whole, may be likened to the movement of a book pulled out from within a row which is meanwhile tipping over. As it slides forward, my book as a crystalline block slips within the row of books. But as they all tip simultaneously, its orientation remains uniform with that of the other books on that shelf during that rotation, and its leaves certainly retain a common orientation, while the book itself tips and while the row of books as a whole is rotating relatively to the books on the other shelves.

In what follows it is most convenient to proceed as if the plastic deformation occurred by crystalline slip alone, and later to ask how far twinning and rotary or fluid movement may replace this slip.

389. Rupture and Fracture.—For clearness let us restrict "*fracture*" to its mineralogical sense of the surface exposed by breaking an object, and let us use "*rupture*" to denote the act of breaking. For instance, the fracture is the path of rupture, the last stage of the path of plastic deformation.

390. The path of deformation and rupture is governed first by the nature of the existing stresses and second by the properties of the material itself. Under the nature of the stresses we may have to distinguish their (1) direction; (2) intensity; and (3) rate of increment, which is very rapid in the case of shock but very slow in quiescent or so-called static testing.

Under the properties of the material itself we may have to distinguish (4) its structure, whether amorphous or crystalline, and whether jointed,

cellular, slaty, etc.; (5) the relative strength of its various internal surfaces of low cohesion, its trans-crystalline cleavages and partings and its intergranular boundaries or joints; and (6) the changes which occur in this relation during deformation.

391. The Common Paths of Deformation.—The paths into which the stresses tend to direct the deformation depend on the stresses themselves, and therefore are common to all kinds of material. This will be elaborated in §401.

In a perfectly isotropic body these common paths might be followed accurately, but in crystalline and other anisotropic bodies the actual path of deformation is deflected more or less from these theoretical directions by the structural peculiarities of the body itself, and in particular by its internal surfaces of weakness which deformation naturally tends to follow.

The actual path of deformation, though deflected thus from the theoretical, may yet belong to the same family. Thus the plane of maximum compressure shearing stress on a column crosses it at an angle of 45° to its axis. Theoretically 45° is the angle at which every column should shear. But a column of cast iron thus crushed shears not at 45° but at a steeper angle, whereas one of wood shears at a flatter angle as in Fig. E of Plate 20, and so the properties of each substance modify the angle at which it shears. Yet this oblique shear remains of the family of the theoretical shear. Paths of deformation belonging to this family, and hence to a family common to themselves, are in this sense "common"[1] to all materials. Hence their name "common paths." Let us glance at a few examples.

If we stretch a strip of India rubber, coat it with stearine, and later allow it to contract, the compression breaks the stearine up into little rhomboids. A plate of glass pierced by a bullet shows a radial and a circular system of lines, as in Fig. A of Plate 20, whereas a plate of gelatine when bent shows oblique conjugate spirals. Pressing a spherical punch against the center of a steel disc while it rests on an annular support sets up the very beautiful pattern shown in Fig. A of Plate 19, with radial lines and logarithmic spirals of two conjugate systems which have the center of the disc for their pole.

Stretching a test piece by means of a bolt passed through its head, as in Fig. B of Plate 19, causes markings of five different geometrical groups, illustrated by the key sketch, Fig. C.[2]

[1] "Banal," Osmond, Fremont, and Cartaud, *Rev. de Metallurgie, Memoires*, 1904, vol. 1, p. 16. "Common" is here used in the same sense as in "common property" or "the commonwealth." It ought not to convey any suggestion of "usual."

[2] These are, in Region *C*, both conjugate spirals and secondary circular arcs concentric with the bolt hole, both representing compressive deformation.

In Regions *A* and *B*, conjugate spirals and also secondary lines normal to the bolt hole, both representing tensile deformation.

In Region *D*, two conjugate systems, prolongations of those which occupy the body of the test piece as shown in Fig. 56.

In Region *E*, lines normal to the chief stresses.

Along the edges of the head, Regions *MNP*, discontinuous conjugate systems of curved lines showing that here too bending occurs.

392. Lines of Lüders.—The paths of plastic deformation are of course internal surfaces. When deformation is pushed to rupture, the fracture itself is such a path, such an internal surface laid bare. But most of our study has to do with deformation not pushed to rupture, and here it is not the internal surfaces of deformation which we see but the traces, intersections, or outcrops of these internal surfaces on the outer surface of the specimen studied. Hence we have to do chiefly with these intersections.

Quite as with the internal surfaces of deformation, so with these lines of deformation, those which are of the same family with the theoretical lines are called "common lines." The geometrical lines caused by moderate deformations are called the "lines of Lüders,"[1] after Lüders of Magdeburg, who first described them fully, though they have since been described even more fully by Hartmann.[2] It is these that we have seen in Plates 19 and 20.

393. Specific Paths of Deformation.—The internal surfaces of weakness of the substance itself may, on the other hand, deflect the path of deformation so far from the theoretical that it is no longer of the same family, and therefore may be called a *specific path* of deformation. In a slaty substance deformation tends to follow the parallel planes of lamination; in a jointed one like brickwork to follow the joints; in a fibrous one like wrought iron or wood to follow the fiber; in a crystalline one like mica or unburnt steel to follow its crystalline planes; in a cellular one to follow the walls of its cells, be they prismatic as in the basaltic or dendritic structure usual in the outer part of steel ingots, or polyhedral but nearly equiaxed as in burnt steel or sandstone.

The crystalline internal surfaces of weakness, the cleavages and partings, form an important part of our study. The noncrystalline ones are considered in §405.

If these internal surfaces are weak enough they may determine the type of fracture to which the deformation finally leads.

Such fractures and the paths of deformation leading up to them are clearly specific, being determined by the specific inherent properties of the material.

It may be possible and extremely important to distinguish in any given fracture, and in any given path of deformation, not only features of the family of the theoretical path, that is, common features, but also specific features resulting from the internal weak surfaces of the material itself, the common path being made up in detail of minute pathlets of the specific type, in short being an aggregate of specific pathlets.

[1] *Dingler's Polytech. Journ.*, 1860, vol. 155, p. 18.

[2] *Congrès Internat. des Methodes d'essai*, July, 1900, in connection with the Paris Universal Exposition, Dunod, Paris, 1901, vol. 1, p. 95. Also Hartmann, "Distribution des déformations dans les métaux soumis a des efforts," Berger-Levrault, Paris, 1896; Fremont, *Bulletin Soc. d'Encouragement*, Sept., 1896; and Hartmann, *idem*, 1897, (5), vol. 2, p. 103.

394. Irregular Paths of Deformation.—When the conditions are such that the deformation is pushed very far, it may become extremely irregular, and we may then have great difficulty in determining whether any given feature of it is common or specific. We shall see foldings and dykes caused, for instance, by the extreme deformation which accompanies punching. It would be very difficult to determine how far these are of the common family and how far specific, quite as it would be with the irregular foldings of the Alps.

395. Even in the Presence of a Marked Macroscopic Specific Structure, Deformation may Integrate into a Lüders Line.—Though in the cases considered in the last section but one the common features of the deformation are completely masked by the specific ones, yet the reverse may be true even in cases in which the specific surfaces of weakness are not only very marked but on a very large scale.

Thus Fig. E of Plate 20 shows how wood, if compressed parallel to its fiber, may yield by shearing along the lines of Lüders at an angle of about 65° with that fiber,[1] in spite of its weakness along its fiber as shown by its ready splitting under the axe. It is interesting here that the Lüders line makes with the direction of the push an angle greater than 45°. There is the suggestion that the failure here is by the buckling of the various fibers. If this is true, then this buckling here integrates into a Lüders line.

So, too, if a brickwork column is relatively fresh when crushed in compression, its rupture follows the relatively weak vertical joints, cleaving those bricks which lie in its way; yet as the mass ages and as its mortar strengthens it shows an inclination to shear along the Lüders lines at an angle of about 45° with the direction of the stress, though even now the vertical mortar joints are certainly much weaker than the bricks themselves thus sheared across.[2]

396. The Specific Microscopic Deformation Lines in Steel Integrate into Lüders Lines.—If such integration is possible in masses like wood and brickwork in which the weakness is very marked along very extensive specific internal surfaces, it may naturally be expected in metal, in which these surfaces are minute.

For instance, though deformation and rupture actually follow the transcrystalline planes of low-carbon steel, because these are weak internal surfaces, yet because these planes if not microscopic are very minute when compared with the mass of the specimen as a whole, and because they change direction from grain to grain, each of these pathlets in effect notches only an extremely minute sawtooth into the lines of Lüders, which but for this would

[1] See also J. B. Johnson, "The Materials of Construction," Wiley, New York, 1897, p. 241, Fig. 113*B*; and Robin, *Rev. de Metallurgie, Memoires*, 1911, vol. 8, p. 454, Fig. 21.

[2] P. H. Bates, U. S. Bureau of Standards, Pittsburgh, Pa., private communication, Nov. 27, 1914; James E. Howard, *The Engineering Record*, 1913, vol. 67, p. 332, "Tests of Two Brick Piers of Unusual Size." The piers were about 4 ft. square and 12 ft. high, and weighed a little over 13 tons each. The picture, taken after crushing, of one of these laid with cement mortar shows strong indications of oblique shearing.

be smooth even microscopically. Indeed, these sawteeth are actually of a lower order of magnitude than the grains themselves, because in crossing a given grain the deformation rarely follows a single cleavage plane but zigzags back and forth from one to another, as shown in Figs. B and D of Plate 20.

It is in this way that the puckering figures which deformation sets up on previously polished steel surfaces, as shown in Figs. 56 and A of Plate 19, though made up of microscopic trans-crystalline straight pathlets, have yet integrated into the geometrical common or Lüders lines, often smoothly curved, as a sequence of short enough tangents integrates into a smooth circular circumference.

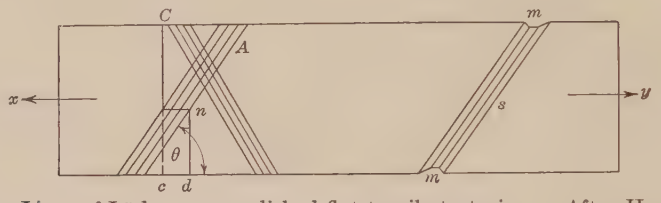


Fig. 56.—Lines of Lüders on a polished flat tensile test-piece. After Hartmann.[1]

397. Scale of the Deviation from Smoothness.—What has just been said applies to very low carbon steel, to Hadfield's manganese steel, and to 25 per cent. nickel steel, for each of these consists essentially of grains of a single substance, alpha iron or ferrite in the first of these cases and austenite or gamma iron in the two others. It should apply to other pure metals and to alloys consisting of a single constituent. In higher carbon steel the conditions are complicated by there being grains of different substances, ferrite and cementite intermixed as pearlite, and either pro-eutectoid ferrite or pro-eutectoid cementite. Moreover, the grains of the pearlitic ferrite are of a lower order of magnitude than those of the pro-eutectoid ferrite of low-carbon steel, and the pearlite itself, though composite, yet often acts as if it were a single substance, so far does it, as a whole, differ from the enclosing pro-eutectoid element.

Because, as we saw in the last section, the scale of the deviations from smoothness is less than that of the individual grain size, this scale should decrease as the grains become smaller, being much smaller in pearlite than in ferrite, and smaller still in sorbite and troostite. Hence comes the coarsening of the fracture caused by overheating, which assembles the ferrite of low-carbon steel into large grains, and converts higher carbon steel into a network with coarse continuous masses of pro-eutectoid ferrite or cementite if the cooling is slow, or, if it is relatively fast, leaves pseudomorphic traces of the coarseness which the austenite reached during the stay above the transformation range (§753). Hence, too, the fracture becomes finer as rising carbon content substitutes the finer pearlite for the coarser ferrite; as grain refining lessens the size of the ferrite network; as we replace pearlite with

[1] "Distribution des deformations," Berger-Levrault, Paris, 1896, Figs. 4 and 5.

the irresoluble sorbite; and as, by increasing the content of the sluggardizing elements, manganese, nickel, etc., we decrease the size which the growth of the microconstituents has time to reach in a given heat treatment.

That the scale of deviation from a strict geometrical following of the directions of stress increases with the coarseness of the granular structure is illustrated by Fig. F of Fig. 20, which shows that a symmetrical wedge of extremely coarse-grained thermit steel, under normal compression applied at its top and bottom, warped greatly out of shape. In this case the whole face of the wedge was made up of only five grains, each plainly visible to the naked eye. In following a direction of easy flow in some one of these great grains, to an extent out of proportion to the actual flow in other directions, the wedge has warped thus strikingly.

398. Example of the Development of Lüders Lines on a Fine-grained Steel Tensile Test Piece.—Hartmann's very extended observations give us a most interesting picture of this development.

If a previously polished and fine-grained steel tensile specimen, such as is shown in Fig. 56, is strained with very gradually increasing stress, the surface remains bright throughout till the elastic limit is approached, when an oblique transverse line A shoots across it so rapidly that the eye cannot follow its progress. With the finger tip we recognize readily that this line is a slight depression. It is the first symptom of permanent set.[1] The metal has yielded by constricting along a plane which makes with the direction of the pull an angle of about 60°. This angle was found by Hartmann to be 63° for spring steel hardened, 58° for the same steel when annealed, and between 53° and 65° for seven other metals and alloys, and thus materially greater than the angle, 45°, which the plane of the maximum shearing stress makes with the direction of pull (§401).

If the stress is increased progressively, this first line may widen, or additional lines either parallel with it, or conjugate[2] with it as at B, may shoot across the polished face in like manner. Each line represents a slight addition to the permanent set. As we strain further, these lines increase in number, at first in rough proportion to the stress, and later more rapidly than

[1] Fremont found deformation lines on a polished tensile test piece of low-carbon steel which had undergone less than 39 per cent. of the stress corresponding to the yield point, or, more accurately, corresponding to that which would have just sufficed to dull the whole surface. That is to say, after a stress of 8.5 kg. per square millimeter or 12,260 lb. per square inch there were some lines present on the previously polished surface, whereas after a stress of 21.5 kg. or 31,000 lb. per square inch, though there were many such lines, yet they did not form a continuous sheet over the whole surface, which happens with such steel at the yield point, according to his observation (*Bulletin Soc. d'Encouragement*, 1903, Part 2, pp. 357 and 364). See also Osmond, Fremont, and Cartaud (*Rev. de Metallurgie, Mem.*, 1904, vol. 1, p. 35), who reasonably interpret this experiment as showing that the local elastic limit may be reached under a stress less than one-third that of yield point, or jog. Kirkaldy's results harmonize with Hartmann's as regards the angle which the Lüders line makes with the direction of pull. He found that, when rupture takes place in a steel tensile test piece under shear, it is along a plane inclined at 65° to the pull. Hodgkinson determined this angle for cast iron in compression as 35° (Ewing, "Encyclopædia Britannica," 1911, vol. 25, p. 1016).

[2] Reciprocally related, in this case related as right and left.

it. No change occurs in their direction, which is thus independent of the intensity of the stress.

The apparent widening of these oblique lines seems to represent the sudden birth of additional lines so near the earlier ones as to suggest their widening rather than their duplication.

A given line such as *A* may change to the conjugate direction *C*, but even in this case it crosses the test piece completely. By a repetition of such changes back and forth the familiar zigzagging of the fracture may be explained.

A line such as *Cc*, drawn across the test piece before the test, becomes stepped as at *n* where it is crossed by these oblique lines, and later curved smoothly by the uneven accumulation of such steppings in the necking part.

Throughout the test piece the various lines vary in depth even where they touch. Parts of the test piece remain free from them. Thus the stretch occurs discontinuously. Hartmann found this general behavior true not only of the pure metals, such as platinum, copper, tin, and lead, but also of composite alloys such as steel, bronze, and brass.

The final necking, *s*, is, according to Hartmann, the sum of a great number of these elementary constrictions in close proximity, as sketched in Fig. 56. This, be it remembered, refers to the behavior of a presumably very fine-grained steel, with grains so small that the path of rupture, in following their cleavages, notches only undetectably short teeth in the Lüders lines. Moreover Hartmann's description applies to conditions selected with the skill of an expert. These lines can be traced to the best advantage very early in the deformation, and even then they are far from prominent under usual conditions.

The arrangement of the lines of deformation in low-carbon steel may be shown also by Chappell's[1] ingenious method of heating the strained and polished test piece to about 700°, when the parts in which the deformation has reached Sauveur's critical stage coarsen, while those which either have not reached it or have passed beyond it do not. On the surface of a test piece treated thus after straining to the yield point, bands appear running at an angle of about 60°, and differing from each other in the coarseness of their granulation.

399. Example of the Development of the Lines of Lüders in a Coarse-grained Steel.—Figs. C and D of Plate 20 show in natural size the progress of the path of rupture in an extremely coarse-grained silicon steel. The coarseness of the large light grain *B* through which much of the fracture passes should be noted. Under an absolute stress of 630 lb., or 18,300 lb. per square inch, a crack opened approximately at 45° to the direction of the pull, and crossed from grain *A* into grain *B* with but little deflection. Under a stress of 1,106 lb. or 32,300 lb. per square inch this same crack lengthened to about ⅞ in., still without material change of direction. Under a stress of

[1] *Journ. Iron and Steel Inst.*, 1914, vol. 89, No. I, p. 492 and Fig. 16, Plate XLIX.

1,300 lb. or 37,900 lb. per square inch rupture became complete, changing in grain *C* to a direction conjugate with its initial path. That a stress of only 18,300 lb. per square inch sufficed to start the first crack, though the section when diminished by the continuation of this crack did not yield till the total stress was more than doubled, indicates that the stress was not evenly distributed.

The most interesting feature is the set of sawteeth shown between letters *B* and *C* in Fig. D of Plate 20, and on a larger scale in Fig. B. These show us how it can result that rupture thus follows a Lüders line which is roughly straight even in crossing a grain boundary. In any given grain this line is composed of sawteeth, which happen here to be coarse enough to be seen. The alternate edges of these teeth follow two crystallographic directions of that grain. In Fig. B of Plate 20 we may distinguish one set as running west and the other east of north. Clearly by varying the ratio of the length of the easterly lines to that of the westerly ones we can give the resultant path any direction between these two extremes, the easterly on one hand and the westerly on the other.

If, as is inferred in §445, slip may occur either along any of the 12 directions of the {211} trapezohedron in any one grain, a resultant of any desired direction can be had by combining these various directions in different pairs, and each pair in different ratios. In short, no matter how the orientation of any individual grain is inclined toward the stress, and toward the Lüders line which that stress tends to develop in the material in hand, a resultant path of slip can form in that grain, conforming in detail with its crystalline planes yet integrating into the natural direction of the Lüders line. So with rupture, whether it follows these 211 planes or the cubic cleavages.

400. Development of Lüders Lines in a Normal Structural Steel.[1]—In the tensile test of a previously polished piece of such a steel, lines in a general way like those described in §415 appear, many of them oblique as in Fig. 56, but some of them normal to the pull. These naturally start near the grips, because here the stress is the least evenly distributed, so that here the local stress at some most severely strained point reaches the elastic limit earlier than mid-length of the test piece. With the advent of these first lines comes the yield point.[2]

As the stress increases new lines appear farther and farther from the grips, so that the dulling of the polish creeps from both ends toward the mid-length of the piece. As the dulled parts meet, the end of the jog or step in the stress-strain diagram is reached, and the curve again starts to rise.

[1] J. B. Johnson, "The Materials of Construction," Wiley, New York, 1897, p. 12, and Plate I.

[2] The apparent discrepancy between Johnson's finding that the first lines appear at the yield point and Fremont's finding them with a stress less than one-third of the yield point is probably due to the higher polish used by Fremont, which would detect deformations wholly invisible with the probably rougher polish of Johnson. Beyond this, Fremont's finding the whole surface covered with lines at the yield point may refer to strain just after the beam has dropped, whereas, Johnson's yield point at which the first lines appear is probably the point at which the beam is about to drop.

401. Deviation of the lines of Lüders from the direction of maximum shearing stress, which makes an angle of 45°[1] with the direction of a longitudinal tensile or compressive stress.

The shearing stress, per unit of a section cut at this angle, is half the direct tensile stress on a cross-section normal to the axis of the test piece.[2] Hence, if the test piece actually sheared across exactly at 45° that would mean that the resistance of the material to shearing is less than half its resistance to a direct pull, per unit of section; if rupture followed a true plane normal to the pull, not zigzagging but smoothly, that would mean the opposite.

The deviation of the direction of the lines of Lüders from 45° in these various metals studied by Hartmann may be interpreted as meaning that these substances yield by preference on a section on which the shearing stress, though not at its maximum, is yet accompanied by a direct pull which is greater per unit of section than it is on the section at 45° at which the shearing stress is at its maximum, and hence that the existence of a greater pull on the former section lessens the resistance there to shearing. So, too, shearing in compression along a plane which makes an angle more acute than 45° with the direction of the push may be interpreted as meaning that here the existence of a push normal to the section increases the resistance to shearing.[3]

To vary the point of view slightly, the fact that the Lüders lines lie neither at 45° along the direction of maximum shearing stress, nor normally to the axis of the test piece along the direction of maximum pull or push, means that the metal offers its least resistance to a combination, in a certain specific ratio, of a direct pull or push with a shearing stress. This may be

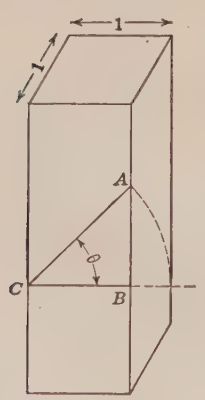


FIG. 57.—The direction of maximum shearing stress is at 45° with the direction of the pull.

[1] That the maximum shearing stress is at an angle of 45° to the direction of the stress is readily seen. Let the body shown in Fig. 57 be a short fragment of a square tie rod of unit cross-section subjected to a pull in the direction of the arrows.

Let P = the pull per unit of section CB.

Let CA = the area of any given section at an angle θ to the cross-section CB, = $CB : \cos\theta$.

If P is resolved by means of a parallelogram of forces into a stress normal to CA, and the shearing stress, P_t, parallel to it,

$P_t = P \sin\theta$.

Then p_t = the shearing stress along CA per unit of area, $= \dfrac{P \sin\theta}{1 : \cos\theta} = P \times \sin\theta \times \cos\theta$, which reaches its maximum, = 0.50, when $\theta = 45°$.

Then $p_t = P \times 0.707 \times 0.707 = P \times 0.50$, so that it is half the direct pull on CB, per unit of section.

By like reasoning the angle of maximum compressive shearing stress is also 45°.

[2] Compare Ewing, "The Strength of Materials," "Encyclopædia Britannica," 1910, 11th edition, vol. 25, pp. 1007, 1008, and 1016.

[3] Navier's interesting interpretation, starting with the premise that the angle of rupture must be that which will cause failure under the least load, is that the resistance along planes of variable inclination is affected by the natural angle of repose of the material. From this he deduces the law that under compression the angle of rupture is 45° plus half the angle of repose (J. B. Johnson, "The Materials of Construction," Wiley, New York, 1897, pp. 25–27).

called the most effective ratio of direct to shearing stress, or briefly the most effective stress ratio.

402. Interpretation of the Geometrical Arrangement of the Lüders Lines.—Because these lines are so strikingly geometrical in the cases which we have seen, we may regard their apparent irregularity, under extreme deformation, as meaning, like the irregularity of the foldings of the earth's crust, simply that our geometry is not subtle enough to recognize their law, remembering that even the human form could probably be expressed by a formula of an extremely refined mathematics.

The directions not only of maximum shearing stress but also of the most effective stress ratio are naturally arranged geometrically. Between the shoulders of the test piece of Fig. B of Plate 19, where the stress is a direct pull, they are parallel, conjugate, and straight; in those parts where there is much bending, they are conjugate, nearly parallel, and curved like an eyelash, but still geometrical.

The geometrical arrangement of the Lüders lines means that microscopic deviations from the line of maximum stress caused by the structural characteristics integrate into a constant or at least a regular whole. This has already been noticed in connection with Fig. 56, in which the angle θ, instead of being 45° and thus coinciding with that of the direction of maximum shearing stress, has been deflected to about 60° or by some 15°. So with the deflection of the lines of rupture in granite, limestone, and other materials from the direction of maximum shearing stress by the specific properties of these substances.

403. Direct and Oblique Lines of Lüders.—In Fig. B of Plate 19 most of the lines are oblique to the direction of the main pull, but those at E, A, B, and C are normal to it, and suggest the pleatings which compression causes in the sides of an accordion or the folds caused by belting a curtain in. In a brickwork column the shear is likely to be vertical and therefore parallel to the stress. In the plate of glass of Fig. A of Plate 20 the cracks are of the direct type, some parallel and some normal to the radial stress.

404. Microscopic Geometrical Structure as Evidence of Crystalhood.—We now see three reasons why the geometrical arrangement of the Lüders lines does not suggest that the metals in which they occur are crystalline, the reasons, first, that they result from the geometrical arrangement of the stresses; second, that they are geometrical in amorphous bodies such as the glass of Fig. A of Plate 20; and third, that when they occur in metals their direction is common throughout a macroscopic region composed of great numbers of distinct crystalline grains, each presumably differing in orientation from its neighbors.

But if the microscopic lines of deformation should prove not only to be geometrically arranged but to have a geometrical arrangement which, though constant in each grain, varies from grain to grain, that would be evidence of crystalline structure. For though there is nothing in the nature of an external stress which can have this specific effect on the arrange-

ment of these lines, this effect is exactly that which the crystalline and hence geometric structure of the grains themselves would naturally cause, because the orientation of the structure is necessarily constant throughout each grain, and presumably varies from grain to grain. This then is the point to which our attention should be directed, the uniformity of orientation of the deformation lines in a given grain and its change from grain to grain.

Some very illustrious investigators have, I believe, been misled by failing to have this specific point in mind. With it before us we shall not be misled for instance by the specific planes of weakness caused by the longitudinal sheets and rods of slag in wrought iron, by slaty cleavages, by the crossing of grain boundaries, by occasional resemblances to the lines of Lüders, etc.

405. Noncrystalline Internal Surfaces of Weakness.—Because we have to distinguish these carefully from the trans-crystalline planes which testify to crystalhood, we may well classify them. They naturally fall into four groups, A to D, differing from each other in the shape of the network which they cause.

(A) The network may be irregularly cylindrical,[1] that is, it may be generated by the irregularly curved movement of a line parallel to its own length, as in the case of fibrous wood, and of the rods and sheets of cinder in wrought-iron bars, and the phosphoric bands and ghosts in steel bars and rails.

(B) It may consist of nearly plane parallel surfaces, as in brickwork, in slates, and in plates and sheets of wrought iron and of banded and ghost-bearing steel.[2]

(C) It may consist of surfaces nearly plane and not parallel to each other or to any line, but lying at all angles, as in the granular structure of metals. The planes of such a network may occasionally lie parallel to three conjugate planes, in which case it gives rise to pseudo-crystalline grains. The pentagonal dodecahedron, so familiar in pyrite, is often simulated in this way.[3]

(D) It may consist of nearly plane surfaces not parallel to themselves, but all parallel to a given axial line, as in the prismatic or basaltic structure. Such basaltic columns may be formed in various ways, and they may approximate the shape of true hexagonal prisms.[4] The outer part of ingots of steel, of white cast iron, and of many other substances very often consist of such columns with their length normal to the cooling surface even if this is vertical, showing that here they are not due to convection currents as they are in certain substances.

Though the columns of basalt itself are not true crystals (§107), the

[1] Geometrically speaking, a cylinder need not be circular in section, but may be generated by a straight line which, remaining parallel to itself, moves through any curved path.

[2] See Oberhoffer, *Zeitschrift anorg. Chemie*, vol. 81, p. 156, and Plate II, Fig. 7.

[3] Osmond, Fremont, and Cartaud, *Rev. de Metallurgie, Mem.*, 1904, vol. 1, p. 12.

[4] For an extremely interesting discussion of the cellular structure see Desch, *Journ. Inst. Metals*, 1914, No. 1, vol. 11, p. 75. See also §§107 and 773 *et seq.* of the present work.

columnar masses in metals certainly often are, in the sense in which the polyhedral grains are crystals, having perfect internal crystalline structure in spite of their irregular outward form. This is shown by the fact that each of these columns etches to a uniform brightness,[1] as shown in Fig. H of Plate 18, and that the relative brightness of each column relatively to its neighbor changes as a whole when the direction of the light is changed. It is also shown by the fact that when a polished surface cut normally to these columns is deformed, certain columns rise above the surface, quite as the grains in a polished surface do, as shown in Fig. C of Plate 21, and by the uniformity of direction of the deformation lines within each column and their change of direction from column to column, as shown in Figs. F of Plate 16 and C of Plate 21. In these micrographs each uniformly oriented area is the section of one of these columns.

This uniformity of structure within each column is masked in carbon steel by the structural metamorphism which arises in cooling through the transformation range. But it is readily seen not only in the manganese steel of Figs. F of Plate 16 and C of Plate 21, which does not undergo this transformations, but also in other solid solutions and in simple metals. Indeed these columns differ from the common grains only in being greatly inequiaxed.

Networks of any of these kinds may exist in either crystalline or amorphous substances, and hence they are no evidence that the substance in which they occur is crystalline. Because their surfaces are often plane, the existence of plane surfaces of weakness, whether shown in the fracture or in polished metallic sections, is not strong evidence of crystalline structure. In order to form such evidence these planes must be geometrically arranged, in parallel groups, making definite angles with one another or with the inferred crystalline axes.

406. Amorphous Smearing of the Metallic Surfaces Prepared for Microscopic Examination.—Because a surface which is to be seen with the microscope must be very nearly plane in order that its different parts may be in focus simultaneously, microscopic examination is confined chiefly to surfaces which are either nearly plane naturally, such as those of crystal faces and cleavages, or are made very nearly plane by grinding and polishing. Unfortunately this polishing coats the surface with an amorphous stratum essentially different in condition from the underlying metal (§508). If we wish to see the true structure of the metal directly, we have to remove this amorphous layer by etching. But in many cases this structure is made to reveal itself indirectly, by means of the specific forms which it causes deformation to take, as for instance in creating slip bands and Neumann bands, which are accompanied by motion enough to cause distinct and easily recognized disturbances of the upper surface of the amorphous layer.

[1] See the Author, "Piping and Segregation in Steel Ingots," *Amer. Inst. Min. Eng.*, 1907, vol. 38, p. 41, Fig. 15.

CHAPTER 16

SPECIFIC DEFORMATION. SLIP BANDS

Of the six types of movement which may contribute to the plastic deformation of metals, let us here consider intergranular movement and the so-called slip bands.

407. Intergranular Movement.—When ferrite, for instance in the form of the very low-carbon A.R.M steel, is plastically deformed, what seem to be the grain outlines appear on a previously polished surface, as shown in Fig. E of Plate 21. On pushing the deformation further, this same uplift is greatly exaggerated, as in Fig. G of Plate 21. Other suggestions of such grain uplift are seen in Figs. C of Plate 21, E and F of Plate 24, B and D of Plate 31, and A, B and E of Plate 32 of ferrite, and Figs. B and C of Plate 21, D of Plate 24, B of Plate 38, and E and G of Plate 39 of austenite.

On copper plating[1] a surface on which this grain uplift has occurred, and then cutting a section across it, the uplift of individual grains is clearly confirmed, as in Figs. D and F of Plate 21 and H of Plate 25.[2] Note in Fig. F of Plate 21 that grain 2 has been pushed up past grain 1, and grain 3 past 4 and slightly past 2. The movement of 3 relatively to its neighbors is seen more clearly in the outline of the black shadow which separates the light ferrite from the dark copper. In Figs. D of Plate 21 and H of Plate 25 grain 2 has in like manner been pushed up past grain 1.

Note carefully that in those cases in which the deformation of the surface is relatively slight, as in Figs. D and E of Plate 21, and H of Plate 25, the movement seems to have been very close to the grain boundary, if not indeed in it, whereas in those cases in which the deformation is greater, as in Fig. G

[1] This copper plating was done by Rosenhain's ingenious method. The profiles of the changes in level which deformation causes on a previously polished surface are disclosed by copper plating that surface, and then cutting, polishing, and etching a section normal to the original polished face and preferably to the length of the slip bands. The black line usually seen between the copper and the steel is a groove or a shadow caused by the difference in level between the steel and copper. Polishing should remove the softer copper faster than the steel, but etching should have the opposite effect (Rosenhain, *Journ. Iron and Steel Inst.*, 1906, No. II, vol. 70, pp. 195–198). An improvement on the earlier procedure is given by Rosenhain and Humfrey, *Journ. Iron and Steel Inst.*, 1913, No. 1, vol. 87, p. 261.

[2] I find a very striking case of grain uplift in Rosenhain's Fig. 7, *Journ. Iron and Steel Inst.*, 1906, No. II, vol. 70, Plate XLI, after p. 224, though I do not remember that attention has been called to it before now. Here the right-hand grain has been pushed up past its left-hand neighbor. Like cases of grain uplift are shown by Howe and Levy, *Trans. Amer. Inst. Min. Eng.*, 1914, vol. 50, Figs. 9, 10, and 11 after p. 536 and like roughening of a previously polished surface is shown by Fremont (*Bull. Soc. d'Encouragement*, 1903, Part 2, Figs. 6 and 7, pp. 357, 358). Galy-Aché pointed out in 1913 that this demarcation of the crystalline grains, "absolutely analogous to those developed by etching," appears on the previously polished surface of a test piece "as soon as the elastic limit" (yield point) "is passed" (*Rev. de Metallurgie, Memoires*, 1913, vol. 10, p. 588).

of Plate 21, while there is like evidence of uplift along some of the grain boundaries yet there is much to suggest that the surface roughening occurs in very large part by the bulging up of the central parts of the grains. Of course any uplift along the boundaries in the early stages of deformation must needs persist through the later stages, so that the impression which we get is that the uplift along the grain boundaries is most marked in the early stages of deformation, and that the bulging up of the grain centers is accented in the later stages.

Other instances of uplift along the boundaries in the early stages of deformation are seen in Figs. F of Plate 24 and A and B of Plate 32 of ferrite, and B of Plate 38 and E of Plate 39 of austenite, and of central bulging in the later stages in Figs. D of Plate 31 of ferrite, and D of Plate 24 and G of Plate 39 of austenite.

These observations and inferences seem to me of the first importance for interpreting the mechanism of deformation, as I shall point out in §§479 and 497.

408. Examination of Certain Cases of Grain Uplift.—In Fig. E of Plate 21 the undulation of the surface has outlined areas so like the grains developed by etching that we interpret them so. Fragments of grain boundaries favorably inclined reflect light into the microscope and hence look bright, while others inclined unfavorably look dark. In Fig. D of Plate 24 the grain relief is still greater.

The inference that these uplifts are of grains is supported by the fact that, when the grains themselves are columnar, the areas thus uplifted on deformation not only have this same columnar shape, but are marked off sharply from each other by just the changes of orientation which habitually occur at grain boundaries. Thus in Fig. C of Plate 21, which shows the markings caused by deformation on the previously polished surface of a casting of manganese steel, certain of the areas marked off from their neighbors by this grain uplift have the columnar shape common in these castings, and are crossed, often completely, by straight parallel slip bands which change direction sharply at the interareal boundaries.

That these columnar crystals are persistent is shown by their appearance in Fig. AA of Plate 21, on repolishing and then deforming a manganese steel casting after heating it to 1,100° and quenching. Here the columnar shape of the uplifts replaces the granular shape seen in Fig. E of Plate 21. The structural persistence should of course be far greater in manganese steel, and indeed in most metals and alloys, than in carbon steel, because they lack the transformation which this steel undergoes every time that it is heated or cooled.

We owe to this transformation our ability to refine the grain of carbon steel by heat alone. It is because other materials lack such a transformation that their grain refining requires the break-up of the grains by plastic deformation, followed often by annealing to remove the brittleness which that deformation causes.

In many cases the grain area is marked out simultaneously by grain uplift and by the stopping of the slip bands or their change of direction, as at the boundary *bd* of Fig. F of Plate 22. But sometimes the uplift is unaccompanied by visible slip bands, as in many parts of Fig. D of Plate 24. Sometimes the existence of a grain boundary which is not shown clearly by grain uplift may be indicated by the arrest or change of direction of slip bands, as along the imaginary lines *abcd* of Fig. C of Plate 24.

One gets the impression that the grain uplift is more clearly marked when the grains themselves are small, as would indeed naturally be the case. Coarseness of grain implies great development of crystalline structure, and extended slip planes along which movement is correspondingly easy. Any path of low resistance tends to divert the internal movement to itself and hence away from other paths, in this case away from the grain boundaries.

409. Suggestions of Grain Uplift Difficult to Interpret.—Having thus interpreted these uplifts as marking out areas which are the sections of true grains, true crystals lacking outward form, we incline naturally to a like interpretation of Figs. A, B and E of Plate 32. Thus the reversed curve *ab* of Fig. B might well be a fragment of a grain boundary, and so might the plateau edge *abc* in the lower left-hand part of Fig. A, and the boundary *abc* of Fig. E, the coarseness of these supposed grains in Figs. A and B corresponding to the enormous grain size of this specimen of silicon steel.

Yet the apparent identity of orientation of the various little peaks or pits, we hardly know which, immediately to the left of the letters *a* to *p* of Fig. E of Plate 24 inclines us to reconsider this interpretation of Figs. A, B and E of Plate 32. Note that each of these little peaks has a steep right-hand and a very flat western slope, and that the edges of all of them run about N-20°-W and N-40°-E. We cannot take each of these for the uplifted corner of a grain, such as we imagined for Figs. A, B and E of Plate 32, because we cannot believe that a pair of sides in each of so many neighboring grains would happen to be parallel. Nor can we refer the similarity of these peaks to their being like markings in a single uniformly oriented grain, because the grains in this specimen are of a lower order of magnitude than this would imply, as is seen in Fig. F of Plate 24, of a neighboring but less sharply deformed part of this same specimen. Their similarity and parallelism through many grains naturally connect themselves with the direction of the stresses rather than with the orientation of the metal, and incline us to call them microscopic lines of Lüders. Yet their shape is not readily referred to fluid movement, and thus we are left without ready explanation of why they are parallel in neighboring grains almost sure to be variously oriented.

So throughout Fig. D of Plate 24 there are short vertical marks very different from the accompanying slip bands, and very nearly parallel throughout the field, though this must contain a great number of variously oriented grains. In many cases, as from *c* to *d*, from *e* to *f*, and from *g* to *h*, these short marks stand in a nearly straight column, and on closer examination

seem to connect themselves with short accordion pleatings of the amorphous flowed surface, irrelated to the orientation of the grains in which they lie.

410. Inadequacy of Block Movement.—Whatever these peaks or pits in Fig. E of Plate 24 and short lines in Fig. D of Plate 24 may be, the traces of block movement about them, whether in the shape of slip bands or of evidence of grain uplift, certainly seem inadequate to account for the deformation of the surface which has occurred. Thus the area about *b'bc* in Fig. E is almost free from slip bands, yet it looks as if it had been greatly deformed. The very abundance of the hairy slip bands in much of this field makes their absence from other and apparently strongly curved parts of it the more surprising. The sharpness of the peaks at the left of *b* and *c* disinclines us to refer the absence of slip bands here to bad focussing. For that matter if these peaks or pits are of block-movement origin, then their two sides should be connected by a base line as at *ac* in Fig. A of Plate 32. Other cases in which rounded foldings are not accompanied by signs of block movement sufficient to account for them are many areas in Fig. D of Plate 24, such as *a* and *b*; the apparently rounded talus of the slope which connects the plateau edge *bc* of Fig. A of Plate 32 with the general level; the hummocks *a*, *b* and *c* of Fig. A of Plate 31; the rounded edges of the steps in Fig. G of Plate 22; and the dying out of the lower ends of the Neumann bands from *c* to *d* of Fig. A of Plate 21, reminding us of the dying out of the curves of the lips on the human cheek.

These cases may of course be explained away later. For instance, what look like the accordion pleatings from *a* to *b* of this last figure may prove to be a series of small Neumann lines like those from *c* to *d* in this same figure, and the apparently rounded foldings may prove on greater magnification to be made up of an extremely great number of slip bands so narrow and so evenly distributed that we do not suspect them here, somewhat as the savage who sees velvet for the first time across a room does not suspect that its apparently smooth surface is made up of many minute steps from thread to thread.

Again, the ruffling of the surface by the deformation itself often gives rise to contrasts in individual photo-micrographs which might well suggest fluid movement. Thus the contrast between the prominence of the slip bands in the upper left-hand part of grain *abc* in Fig. B of Plate 31 and their absence from the region below the line *bc* at first suggests that the slip so prominent in *abc* is replaced in the area *dbcef* by fluid movement, a suggestion which we set aside when we notice that in the lower part of *abc* the absence of slip bands is almost as marked as in *dbcef*, and may be due to local bad focussing.[1] The amorphous layer may contain or conceal lines.

Nevertheless, till these surprises are explained away, their existence inclines us to admit as a hypothesis worthy of consideration that block movement may be accompanied by fluid movement. Heyn's[2] assertion that the

[1] The reader unaccustomed to photo-micrography has to be cautioned that the inevitable undulations of the surface, a necessary consequence of the deformation itself, are often an absolute bar to having any large fraction of the field simultaneously in focus under large magnification.

[2] Martens-Heyn, "Materialenkunde," Springer, Berlin, 1912, p. 223.

Lüders lines appear earlier than the slip bands on polished tensile test pieces points rather in this same direction, though not very persuasively.

411. Fluid Foldings.—Certain foldings occur which suggest fluid movement strongly. Among these are the accordion pleatings, which are shown in their incipiency from *a* to *b* in Fig. A of Plate 21 and in their fullness in Fig. B of that plate, in neither of which is there clear indication of block movement on this magnification.

412. Slip bands are supposed to be extremly minute microscopic steps, like $a'b'c'$, etc., in Fig. 64 (§455), set up on the previously polished surface of a metal by bending or otherwise deforming it permanently. In the simplest case the treads of these steps represent successive fractions of the initial surface, and the risers represent parts of the slipping planes laid bare by the slip. Under vertical light these treads are bright, whereas the risers are black (§467), though a narrow light band often runs beside the black (Figs. E and F of Plate 22, A, D, E and F of Plate 23, E, F and K of Plate 38, and D and E of Plate 39), for which an explanation is suggested in §471. Indeed, the bright band is sometimes more prominent than the black of the riser. According to the theory of Ewing and Rosenhain, now generally accepted, each slip band represents the slipping of a crystalline stratum or slice past its neighbor along a crystalline slip plane, like the little step at the edge of each card of a pack slid slightly out of square.

The usual appearance of slip bands in copper is shown by Figs C, F, I and L of Plate 38, and of those in low-carbon steel by Figs. E and F of Plate 24 on small magnification and by Figs. F of Plate 22, A to F of Plate 23, and B of Plate 31 on large magnification. They are of all degrees of straightness and regularity, from those of the copper of Fig. E of Plate 45, of the ferrite of Fig. D of Plate 22, and of the austenite of Figs. B and D of Plate 29 and E of Plate 39, to the very irregular ones in thermit steel shown in Figs. C and E of Plate 22. Extremely straight ones in lead have been shown.[1]

The life history of slip bands will be described in this chapter, and their nature will be discussed in Chapter 18. The slip bands in ferrite are referred to throughout, except where the contrary is indicated. The means of distinguishing them from Neumann bands which often accompany them will be described in §588.

The slip bands in manganese steel will be discussed in Chapter 26, but they will also be referred to in passing in the present chapter.

413. Progress of the Development of Slip Bands in Ferrite under Compression. Definitions.—In a given grain the slip bands usually have a *dominant direction*, which may be followed closely, as when the slip bands are both closely parallel and nearly straight, or may be deviated from materially. In this deviation the slip bands may remain nearly parallel, in which case the deviation is of the whole system of slip bands, a *systematic deviation;* or they may deviate independently of each other, giving us *individual deviations.*

[1] See Ewing and Rosenhain, *Phil. Trans. Royal Soc.*, 1900, A, vol. 193, Figs. 23 and 24 following p. 375.

By the *terminal grain boundaries* I mean those toward which the dominant direction of the slip bands runs; by the lateral ones I mean those along which it runs. Thus in Figs. 58 *A* and *B* boundaries 2, 3, 5 and 6 are terminal and 1 and 4 are lateral. Though there are cases in which this distinction cannot be made rigorously, yet it helps the description which I will now give in §§414 to 417 of the development of slip bands as I have observed it in ferrite under gradually increasing compression in a polished wedge.[1]

414. The First Appearance of Slip Bands.—First a single set of at least roughly parallel slip bands is seen in a few of the grains, as in grain I of Fig. F of Plate 23, and I and II of Fig. A, the neighboring grains remaining free from them.

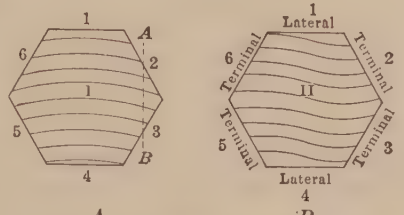


FIG. 58.—*A*, Saucer type of buckling; *B*, *S*-type of buckling.

The appearance of the slip bands at an early stage of their formation, that is to say in the lower part of the wedge, is shown in Fig. 60.

I. Out of a total number of 59 grains examined at this stage, 29 had the slip bands as shown in I of this figure, that is, starting from a point on one side of the middle of the grain and extending across the middle nearly or quite to the opposite terminal boundary.

II. In 14 grains they reached from a terminal boundary less than half-way toward the opposite one, as in II.

III. In 5 and possibly in 10 grains they reached from midway across the grain part way toward both terminal boundaries, as in III.

[1] A convenient way of developing the successive stages in the progress of such deformations with a view to studying them is to polish a truncated wedge on one of its trapezoidal faces, Fig. 59*A*, and then compress it till nearly the whole of that face roughens. We thus preserve every stage in the deformation from the least detectable near the base, to the greatest near the top, and our study can then run the gamut of the various stages in any desired way till we are satisfied. This method has been borrowed and modified from Fremont, "Mesure de la limite elastique des metaux," *Bull. Soc. d'Encouragement Indust. Nationale*, 1903, Part 2, p. 362. It was followed by Osmond, Fremont, and Cartaud in their classical investigation, "Les modes de deformation," *Rev. Metallurgie, Mem.*, 1904, vol. 1, p. 27. These investigators used a truncated pyramid, but for simplicity I have replaced this with a truncated wedge sometimes ½ in. high from base to truncated top, ⅜ in. thick, ½ in. by ⅜ in. at the base, and ¼ in. by ⅜ in. at the top, Fig. 59*A*, and sometimes ¾ in. high, ⅜ in. thick, ¾ by ⅜ in. at the base and ⅜ in. square at the top.

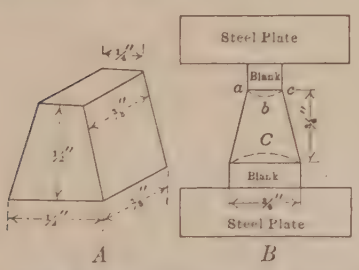


FIG. 59.—*A*, Wedge; *B*, wedge and blanks.

In order to lessen the effects of friction between the wedge and the greased pressure plates, I set a blank, made of material similar to that of the wedge, at each end of the wedge, of the same cross-section as the end of the wedge which it protects, as sketched in Fig. 59*B*. If this precaution is omitted, then the band *abc* running along the top of the wedge remains almost free from deformation, in spite of the greased contact between the wedge and the steel compression plates. Again, below the region where the dulling of the polished surface ceases, and near the base of the wedge, sometimes comes a new region of slight deformation, the segment of a circle of long radius, sketched at *C* in Fig. 59*B*, probably because the top and bottom of the wedge are not strictly parallel.

IV. In 3 grains their arrangement was that shown in IV.

In III the slip bands must have originated near the middle of the grain. In II they cannot have, but must have originated either at or near the grain boundary. In the most important class, I, they may have originated near the middle of the grain and spread thence farther toward one terminal boundary than toward the other, or they may have originated at or near a terminal boundary and spread thence toward the middle of the grain.

Cases in which the slip bands have not reached the terminal boundaries are shown in grain I and at *c* in grain II of Fig. A of Plate 23, and in grains III of Fig. E and I of Fig. F. Cases in which they reach those boundaries are shown in grains I, II, III and IV of Figs. C and D. Slip bands, each of which extends only a short way from a terminal boundary, are shown at *a* and *b* in grain II of Fig. A.

Humfrey[1] finds that the first slip bands, those which appear immediately after passing the elastic limit, "are confined almost entirely to the central parts of the crystals, and only spread gradually toward the boundaries as the straining becomes severe." My own observations, which differ from his in representing compression instead of tension, certainly do not justify so strong a statement, and seem rather to suggest that the slip bands more often start at or near a grain boundary than elsewhere. Again, the slip bands which form during reversals of stress, as in Figs. A and B of Plate 25, though they sometimes give evidence of starting near the grain centers, yet in other cases, as in grains 3 and 6 of Fig. B, rather suggest having started near the grain boundary. This refers to the position of slip bands within the grain in which they originate. Of course those which propagate into a given grain from a neighboring one are likely to enter through the grain boundary.

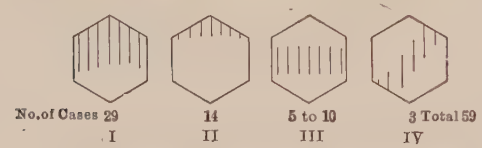


FIG. 60.—Ways in which slip bands originate.

415. Their Development.—Later they appear in more and more of the grains of the wedge under compression, and finally in all of them, increasing the while in any given grain both in individual width and in number, though they may simultaneously become less conspicuous, as their very abundance breaks up the treads or broad white spaces between them with which they contrasted zebrawise so strikingly at first, and as the increasing curvature of the surface makes focussing increasingly difficult. In parts of grains III and IV of Fig. C of Plate 23 the white strips between slip bands have already become so narrow as to suggest that their further narrowing might make them hard to recognize, especially under smaller magnification.

The slip bands formed by repetitive stress may thus run together into irresoluble black areas, such as those in grains 3 and 7 of Fig. B of Plate 25.

With the accumulation of these slip bands comes more or less inter-

[1] *Carnegie Memoirs, Iron and Steel Inst.*, 1913, vol. V, p. 91.

granular upheaval, and these two movements coöperate to dull the initial polish of the test piece.

Long before slip bands can be seen in all the grains they extend completely across some of them, as for instance grains I and II of Fig. E of Plate 23, and in some cases across as many as four grains tandem, as from grains I to IV of Fig. D of Plate 23, sometimes beginning to curve as if for this purpose long before reaching the boundary, as in grain II of Fig. E of Plate 23.

416. The direction of the earliest slip bands, near the base of the wedge, is nearly normal to the stress, that is to the axis of the wedge, though very soon they appear at about 45° to that stress in neighboring grains Then others appear at intermediate angles, though 90° and 45° to that stress seem preferred, even under great deformation. Only when the deformation is marked do they make an angle less than 45° with the stress, and even then they are only very rarely parallel to it.

417. A second set of slip bands can be seen in some grains, as in the left-hand part of Fig. F of Plate 22 and in Fig. B of Plate 23, but this is usually only after deformation is well advanced, indeed very often after it has become so great that only one of two sets can be seen clearly at a time, the other being necessarily out of focus or unfavorably illuminated. These conditions, which may also prevent our seeing simultaneously the whole length of any set of slip bands, explain why two sets of slip bands are rarely seen in a single grain in the published micrographs of ferrite deformed by simple tension or compression, though they often are in those of copper, lead, cadmium, and silver, in which the surface upheaval is so much less for given development of slip bands that three or even more sets of them may be visible simultaneously. In manganese steel, too, more than one set of slip bands is common, as in Figs. C of Plate 21, B and H of Plate 38, and B, D and G of Plate 39. Four sets of slip bands are seen in Fig. H of Plate 21, of the iron nickel alloy called invar. Three, forming an equilateral triangle, are seen in Fig. E of Plate 45 of a crystal of native copper.

There is no fixed relation between the degree of deformation and the appearance of a second set of slip bands in any given ferrite grain, for though many greatly deformed grains have one set only, yet some which lie near the base of the wedge have two sets when they are so slightly deformed that their total number of slip bands is very small. This is true of grain II of Fig. A of Plate 23.

Occasionally both of two sets of slip bands may fail to pass beyond their intersection, thus giving rise to the marking which so often suggests twinning in manganese steel, as at *ab* in Fig. D of Plate 39.

Under reversals of stress, as in fatigue testing, the surface upheaval is so much less than in tensile and compressive testing that two well-marked sets of slip bands may be prominent simultaneously, as in grain 3 of Fig. C of Plate 25.

418. The profiles of slip bands are seen in the cross-microsections Figs. E to I of Plate 25. That which comes much the nearest to the ideal

sketch of Fig. 64 is Rosenhain's Fig. E of this plate. Here the depth of the smallest steps is not more than 0.000,02 in., or of the order of magnitude of a wave of light. In rare cases, in which the deformation has not been severe, the depth may be as much as 0.000,1 in., following the general principle that as the deformation increases the number of steps increases but their depth decreases, indicating that a given step breaks up, with progressing deformation, into a number of steplets. Under favorable conditions Rosenhain finds that the angle of the steps changes on passing from grain to grain.

The appearance of these profiles and their optical behavior will be taken up in §464.

419. The straightness of slip bands varies very greatly even in the same material, and even in the same microscopic field. Those in 25 per cent. nickel steel, in manganese steel, and in copper may often give the impression of perfect straightness, as shown in Figs. F of Plate 16, C of Plate 21, B, D, E and G of Plate 39 and Fig. E of Plate 45, but on closer examination they usually show more or less curvature. Very straight ones have been shown in soft metals such as lead and cadmium, but even in lead they may be very irregular. Their irregularity is greater than would be inferred from published micrographs, which often seem to have been selected so as to show typical rather than usual forms.

In alpha iron or ferrite they are less straight than in most other metals, yet even here they may be nearly straight, as in Fig. D of Plate 22; in area *abd* of Fig. F of Plate 22, in parts of grains I of Fig. A of Plate 23, I, II, III, and IV of Fig. C of Plate 23, II of Fig. D of Plate 23, I and III of Fig. E of Plate 23; and Fig. B of Plate 24.

A fairer idea of their usual degree of straightness in ferrite is given by the fine hair lines in Figs. E and F of Plate 24, and C of Plate 33, and by the broader bands of grain I of Fig. F of Plate 23. Cases of great irregularity are given in Figs. C and E of Plate 22 and B of Plate 35. The first suggests that it represents a section nearly parallel to the slipping planes. In the second there are indeed two dominant directions, about N-20°-E and N-10°-W, which tend to divide the field into rhomboids, but the deviations from these two directions are frequent and marked. Here the extreme waviness of the slip bands certainly gives no suggestion of a geometrical pattern.

They are apt to be the straightest in the early stages of deformation, tending to become more and more irregular as the deformation proceeds. This is true of other metals as well as of ferrite (see §498). On continuing the deformation of the specimen shown in Fig. E of Plate 45 the slip bands changed to a very regular reversed curve.

420. Direction of Systematic Curvings.—When a truncated wedge of ferrite is compressed so as to develop slip bands, in nearly all of the many grains in which they curve systematically the curve which they follow is smooth and suggests the buckling of the leaves of an open book under pressure applied at the top and bottom of the pages. This buckling is usually of

the flat saucer type shown in Fig. 58*A*, but sometimes of the S type shown in Fig. 58*B*, and more rarely fan or broom shaped.

The curvature of the slip bands which any of these forms of buckling causes is usually such that at their ends they curl away from their dominant direction, and thus tend toward becoming parallel to an imaginary line, such as *AB*, Fig. 58*A*, normal to that direction, and hence toward becoming parallel to any terminal boundary which may happen to be parallel to *AB*. But because a terminal boundary may make any sort of angle with the dominant direction and hence with *AB*, the approach of the slip bands to it may progressively either increase or decrease their parallelism to it, increasing that parallelism, for instance, at terminal boundaries 2 and 6 of Fig. 58*A* and 3 and 6 of Fig. 58*B*, and decreasing it at boundaries 3 and 5 of the former and 2 and 5 of the latter figure.[1]

The curvature of the slip bands is in some cases much greater than could be explained readily by any probable shortening or narrowing by lateral compression of the grain in which they occur, and rather suggests that it is a means of enabling slip to propagate across a terminal boundary without having to make any sharp turn (§425).

421. Under Tensile Stress.—If we watch with a microscope the previously polished surface *ABC* of a tensile test piece, Fig. 63, p. 336, we detect no change, according to Rosenhain, till the stress reaches the yield point, when we see in certain of the grains systems of roughly parallel black lines, nearly normal to the pull.[2] With further increase of stress additional systems of parallel lines appear in other grains, departing more and more from the direction of the pull. Then in some of the grains there appear a second, a third, and even a fourth such system, each intersecting the others.

422. Slip bands under repeated reversals of stress[3] are like those under simple tension or compression, in starting as extremely narrow bands running part way across certain grains, and apparently originating preferably near the grain centers, though slip, in propagating into a second grain from one in which it has originated, may well cause a slip band which, as regards

[1] The assertions in this and the preceding section do not really conflict with what Humfrey probably intended to imply (*Iron and Steel Inst., Carnegie Memoirs*, 1913, vol. 5, p. 91), though they lead to a different interpretation as shown in §414. He reports that the slip bands in their incipiency are almost wholly in the central part of the grains, spread gradually toward the boundaries as the deformation increases, continue without change of direction close to boundaries to which they are parallel (lateral boundaries), but narrow and grow less nearly normal to a boundary which they approach at an angle (terminal boundaries).

[2] That these slip bands occur simultaneously with the first permanent set was shown by Rosenhain. Sometimes a single grain remains undistorted by the stress after the others are distorted visibly. On releasing the stress in these cases, actual measurement of the grains in which there are no slip bands shows that they are undeformed, and that the amount of deformation in the others is roughly proportional to the number of slip bands (Rosenhain, *Journ. Iron and Steel Inst.*, 1906, No. II, vol. 70, p. 190, and Figs. 1 and 2).

[3] This section is based chiefly on the work of Ewing and Humfrey, "The Fracture of Metals under Repeated Alternations of Stress" (*Phil. Trans. Royal Soc.*, 1903, A, vol. 200, p. 241), and of Stanton and Bairstow, "On the Resistance of Iron and Steel to Reversals of Direct Stress" (*Proc. Inst. Civil Engineers*, session 1905–6).

the grain thus entered, starts from the grain boundary. But they differ from them first in being straighter; second, in developing chiefly through the widening of certain lines or groups into cracks, with such moderate increase in their number that even at rupture many grains remain free from them; and third, in being accompanied by only very slight general upheaval of the surface, the upheaval caused by any one push being naturally effaced in the following pull. The slightness of the upheaval removes the difficulty of focussing, and hence two intersecting and independent sets of slip bands may readily be prominent simultaneously,[1] as in grain 3 of Fig. C of Plate 25.

In Figs. A to D of Plate 25, which show the same field after increasing numbers of reversals of great stress (32,030 lb. or 14.3 tons per square inch),[2] note the straightness of all the slip bands in all the grains; the contrast between their progressive widening from stage to stage and the very slight increase in their number between Figs. B and C, for instance in grains 1, 2, 3, 4, 5 and 6 of Figs. B and C; the very great number of grains, for instance 9, 10 and 11 of Fig. C, wholly free from them in this latter stage, though 85.7 per cent. of the number of reversals needed to cause rupture had already occurred; and the lack of indications of increasing grain uplift, in the way either of sharper marking of the grain boundaries or of hill and valley building.

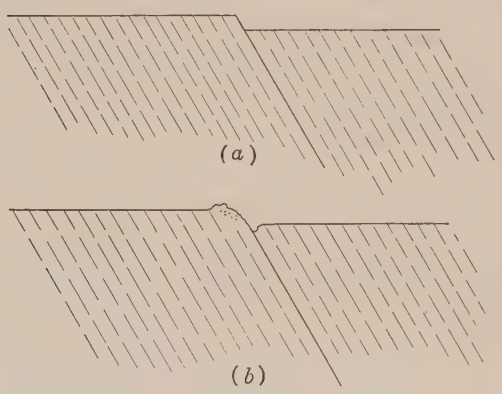


Fig. 61.—The burr formed at the slip band under repetitive alternations of stress. Ewing and Humfrey.[3]

Note further that even in Fig. B, after only about half the number of reversals which at last caused rupture, very many slip bands have lost their initial sharpness which under simple stress persists, and have run together into black patches, for instance those in grains 3 and 7, which even after 30,000 more reversals, as in Fig. D, do not develop into actual cracks.

The slip bands probably started in the middle of grain 8, but in grains 5 of Fig. A, and 3, 5 and 6 of Fig. B they seem rather to have started near a terminal boundary.

On reversal and re-reversal of stress the slip continues by preference, we may imagine, as a repeated sliding back and forth along that fraction of

[1] See, for instance, Figs. 15 and 16 of Stanton and Bairstow, "The Resistance of Materials to Impact," *Engineering*, 1908, vol. 86, p. 735. In this case the fatigue was brought about by a series of impacts alternately in direct tension and compression.

[2] This test was made by Ewing and Humfrey on coarsened, very pure, and uniform Swedish iron, with tensile strength 52,860 lb. per square inch (23.6 tons); proportionality limit, 26,880 to 29,120 lb. per square inch (12 to 13 tons); and yield point, 31,580 lb. per square inch (14.1 tons). The test piece was a cantilever rotating under transverse load. The stresses recorded are the maximum fiber stresses.

[3] *Trans. Royal Soc.*, vol. 200, A, 1903, Fig. 13, p. 246.

that unfavorably placed plane where it has started, and naturally spreads so as both to lengthen and to widen the slip band by which it is disclosed. With the progress of this growth next comes the formation of a burr, shown in Fig. 61, suggesting the heaping up of débris or the welt on the inner upper corner of a rail head, and with it the substitution of a blackened area such as are seen in Figs. B and C of Plate 25 for the individual slip bands; then the starting of cracks, each within a grain, and their running together till, almost immediately before rupture, they stretch continuously across the specimen.

These cracks become prominent when we efface both the single and the grouped slip bands by repolishing and re-etching, as at *AA* in Fig. D, photographed after rupture had occurred in another field.

423. Examples of the Development of Slip Bands under Reversals of Stress.—To illustrate further the behavior of this same metal, when the fiber stress was 11,870 lb. (5.3 tons) per square inch no slip bands appeared after 1,000,000 reversals; when it was 15,460 lb. (6.9 tons) a single slip band was seen after 3,000,000 reversals, running from the boundary of one grain to beyond its middle; when it was locally 16,350 lb. (7.3 tons), slip bands appeared and broadened, suggesting that they might in time cause rupture, though the stress was only 52 per cent. of the yield point and 58 per cent. of the proportionality limit; and when it was 20,160 lb. (9 tons) or more it invariably led to rupture.

If we admit that the formation of the single slip band under a fiber stress of 15,460 lb. per square inch implies that the local elastic limit has been passed, this implies that the elastic limit may locally be only 49 per cent. of the yield point and 55 per cent. of the proportionality limit of the specimen as a whole, as found in the usual tensile test.

Again, under a fiber stress of 27,780 lb. (12.4 tons) per square inch several extremely faint but long slip bands could be seen after the 1,000th reversal; after the 2,000th they were more distinct, and new and much more prominent ones ran a short way centerward from their terminal, or in this case initial, boundary; after the 10,000th the new ones had broadened in their initial part and extended thence faintly across the grain; after the 40,000th they had broadened and heaped up their burr; but they changed little during the following 120,000 reversals, which broke the specimen elsewhere.

In another material certain slip bands showed no change between the 150,000th reversal of direct tensile and compressive stress and the 614,000th reversal which caused rupture, as if the initial unfavorable disposition of these early slipping planes was corrected by some adjustment of the stress through slip.

Not only may the earliest slip bands fail to keep their lead and thus in time to cause rupture, but it may be impossible to foretell till very shortly before rupture which set of slip bands will develop into the fatal crack.

It is in harmony with the slightness of general deformation that rupture

occurs without material change of dimensions of the test piece, as it does in the propagation of a crack under other conditions.

The smaller the fiber stress the fewer the slip bands which form in any one grain, but the larger the proportion of these that propagate so as to form cracks.

Once a crack has started, further change is confined chiefly to its immediate neighborhood, and probably to its propagation, as is the case with cracks in general. That is to say, the stress henceforth is relieved so far by the yielding of the metal at the ends of the crack that little is available either to start new slip bands or widen the existing ones.

A crack may change direction sharply, as if from one set of slip bands to a conjugate one.

424. Discontinuous slip bands, or markings like slip bands, are seen in Fig. F of Plate 33, which shows the deformation set up on a polished surface of coarsely crystalline silicon steel. The specimen was hammered, polished, etched, and then squeezed in a vise. The broad parallel-sided band *MM'* crossing the micrograph diagonally is like a Neumann band. The discontinuous ripple marks, making an angle of about 70° with the horizon, seem to be slip bands,[1] perhaps representing an early stage in their development, to be followed by their extending so as to become continuous. On smaller magnification, as in Fig. D of Plate 33, this discontinuity may escape notice.

In many of the published illustrations there are, both in iron and in other metals, slip bands which show such discontinuities, though they may be less marked than those of Fig. F. These markings are such as might be made either by fluid motion or by the varying slipping of the blocks of a crystalline stratum, some blocks slipping appreciably and thus causing such discontinuous marks, and others inappreciably.

425. Their Behavior at Obstacles; at Grain Boundaries.—They very often stop short at true grain boundaries as in Figs. F of Plate 22, C of Plate 24, and B of Plate 31, as distinguished from those of the second order (§359) which they usually cross, as in Fig. B of Plate 24. Yet they often cross even true grain boundaries in ferrite, curving round, in case their dominant direction is not the same on both sides of the boundary, as if to avoid an abrupt bend at the boundary itself[2] (§420). The crossing of grain boundaries without marked deflection occurs between grains I, II, and III of Fig. D and between grains I and II of Fig. E of Plate 23.

[1] They may possibly be the traces of old scratches which, though wholly removed by the polishing, have yet left altered layers below competent to cause these ripple marks on later deformation. But they are neither straight enough nor continuous enough to be explained readily thus, because scratches are usually straight on this magnification.

[2] The mere fact that a given slip band on one side of a boundary matches one on the opposite side should be accepted only very cautiously as evidence that slip here has crossed the boundary, because such a matching would occur as a matter of chance in a certain proportion of cases.

The apparent angle which the slip bands in adjoining grains make with each other must needs vary greatly with the direction of the plane of the section, as is shown by a very simple experiment, which I commend as helping to grasp these conditions. Cut out of a sheet of stiff paper an isos-

Their passage across certain apparent grain boundaries, caused not by etching but by deforming a polished surface, as in Fig. F of Plate 24, must be accepted with reserve, because these apparent grains may not be true grains, but only fragments of grains broken by the deformation.

426. Mixed Orientation at Grain Boundaries.—Along the grain boundaries there is, in Rosenhain's opinion, a narrow band in which the orientation is a mixture of that of the two adjoining grains, as if dendrites here interlocked, and in Humfrey's belief a region of progressive confusion of orientation (§§700 *et seq.*). To decide whether this is true would need very precise observations directed especially to this point. The forking which the slip bands at times undergo on nearing the boundary is inconclusive, because it extends over a wider area than can be imagined for such a mixture of orientation.

Along many of the apparent grains indicated not by etching but by deforming an unetched polished surface, there is a narrow band which might represent this mixture. Note the light selvage from *a* to *b* in the upper right-hand corner of Fig. D of Plate 22, and along the apparent grain boundaries *ab*, *bd*, and *bc*, of Fig. F of Plate 22. The slip bands do not seem to enter this selvage.[1]

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celes-triangle arrow head, the two sides of which meet at an angle of about 70° like the lines on either side of the intersection *ab* of Fig. H of Plate 39. Set an open book on end on your desk, insert the point of the arrow head, with its base horizontal, between the leaves, and open or shut the book till the arrow head just fits the opening between its leaves. It the bisectrix of the triangle is held horizontal, the leaves will of course fit its sides when the book is opened at this same angle, 70°. If the book is now opened gradually wider and wider, we can continue to make the sides of the arrow fit the space between the leaves by simply raising its point and lowering its base, theoretically till the book is open wide so that its two covers are in the same plane.

Now let the arrow head represent the plane of our microsection, and the leaves of the book represent the direction of two slip planes which approach each other across adjoining grains, and intersect at the grain boundary, the back of the book. Let the book be nearly wide open, the angle between its covers being 170°. The sides of an arrow which, if held horizontally, is to fit between these leaves would have to meet at this same obtuse angle, 170°. If now the arrow is inclined more and more by raising its point while keeping its base horizontal, it will have to be trimmed off so as to be less obtuse, then so as to be a right angle, and then to a more and more acute angle, in order to remain always just fitting in between the leaves, which remain at an angle of 170°. If we replace the arrow by the plane of our microsection, then if we incline it more and more from the horizontal, the angle at which the leaves will cut it will in like manner become first less obtuse, then a right angle, and then a more and more acute angle. In short, the apparent angle between the slip planes on the two sides of a grain boundary varies greatly with the inclination of the plane of the microsection to those planes. Thus in two adjoining grains *A* and *B*, which in fact are within 10° of being parallel, so that slip along one of them in grain *A* might easily cause slip along the other in grain *B*, if cut by a microsection which is strongly inclined to both may seem to make a very sharp angle, so that the continuation of slip bands across the grain boundary looks as if the direction of slip changed by say 70° as at *ab* in Fig. H of Plate 39, though in fact it changes by only 10°.

[1] Such a selvage may arise in etched sections as an optical illusion. If one of two adjoining grains etches as a whole less deeply than its neighbor, so that its general surface stands higher than that neighbor's, then the salient edge of the former is likely to be etched away so as to form a sort of slope down toward the lower surface of the neighbor, and this slope, as seen from above through the microscope, will look like a selvage (Heyn, *Journal of the Iron and Steel Institute*, 1904, No. I, vol. 65, p. 378, and "Mikroskopische Untersuchungen an tiefgeätzen Eisenschliffen," Berlin, 1898). But this etching effect hardly explains the selvage between two grains disclosed

Those cases in which the slip bands are continuous across the grain boundary do not oppose this idea, because they might continue thus across such a region of mixed or of no orientation, as a crack running along the cleavage of a mass of slate would naturally cut straight across any thin transverse parting, no matter how its particles were oriented.

At best this region, if it exists, must be extremely narrow. For instance, along the grain boundary from *c* to *d* at the lower left-hand corner of Fig. D of Plate 22, any admissible region of mixed orientation cannot be of an order of magnitude higher than that of the waves of light, and its width must be a negligibly small fraction of that of the grain at the right of this boundary. Indeed so competent an observer as Heyn[1] denies the existence of such a region in ferrite or in copper not in the cast state. This will be discussed in §§718 *et seq.*

427. On crossing twins and Neumann bands, which are probably mechanical and usually narrow twins, the slip bands may zigzag, or disappear, or change the strength of the contrast between their steps and risers, or, if there are two sets of slip bands, their relative prominence may be reversed (§§558, 600).

It is in harmony with these observations that the slip bands in tin zigzag on crossing the narrow mechanical twins or Neumann lamellæ (§602).

428. At Other Slip Bands.—A slip band probably always faults an older one in crossing it, but the displacement is usually too small to be detected as confidently as in Fig. B of plate 23. The younger slip band should be continuous across the intersection, the older one should not. Thus at the intersections *a*, *b*, *d*, and *e* of Fig. I of Plate 38 the *N*-40°-*W* lines are continuous and the nearly horizontal ones are broken, whereas at *c*, *f*, and *g* the converse is true. In other cases both bands seem to continue across their intersection, as at *a*, Fig. H of the same plate. The undulations so common in these slip bands seem on the whole more marked at their intersections, as at *b*, *c*, *d*, and *e*, in Fig. H. These two micrographs, Figs. H and I of Plate 38 represent manganese steel and copper respectively.

429. At dendrite boundaries the slip bands of manganese steel may cross without sign of deflection, as in Fig. G of Plate 39.

430. The Persistency of Slip Bands.—Though the slip bands themselves are surface markings and hence must be effaced on repolishing, yet if they represent slip along crystalline planes, the new surface formed by the repolishing should intersect these planes very close to where the slip bands were, so that an appropriate etching should cause them to reappear as "slip traces." Bands which may represent these slip traces are indeed

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by the absence of slip bands on unetched polished surfaces, because if the surfaces of the two grains are so nearly on the same level that slip bands can be seen in both simultaneously, then those slip bands should appear on this slope also. The fact that the selvage is not dark shows that any such slope must be extremely flat, for otherwise it would fail to reflect the light into the microscope, and hence would be black.

[1] *Journ. Iron and Steel Inst.*, 1904, No. I, vol. 65, pp. 376 and 378.

found on thus repolishing and etching under certain conditions. These I call X bands, because their nature is as yet unknown. They are described in §§624 to 626 and their nature is discussed in §§634 and 635.

431. The tendency to form slip bands decreases at high temperatures, as is indicated by the observations recorded in §§688 and 698, intergranular motion then taking the place of the trans-crystalline slip by which deformation is accomplished in the cold.

CHAPTER 17

SILHOUETTES

432. Introduction.—In this chapter I consider the crystallographic relations of the slip bands, especially as disclosed by the data of Osmond and Cartaud. From these I infer that though the slip planes in austenite are octahedral, those in ferrite resemble the Neumann lamellæ in following the planes of the {211} trapezohedron. In order that the reasoning may be followed the more readily I try to give, in Chapter 13, in this chapter, and in Appendix I, the needed slight familiarity with a few of the simpler isometric crystalline forms.

433. The Silhouettes, or Compression Figures.—Press the end of your penholder down upon your rubber, and you make a depression of which the outline or *silhouette* is truly circular. Replace the penholder with a square rod and the silhouette becomes square, though with rounded corners. Because the rubber is isotropic, that is equally strong in all directions, the shape of the silhouette reflects only the shape of the instrument which makes it. But replace the rubber with a substance which is crystalline, and therefore geometrically anisotropic, its strength varying with the direction, and the silhouette reflects this geometrical inequality of strength, by extending farther from the point of the depressing instrument in the direction in which the strength is the least than in other directions, and thus forming a geometrical figure.

434. Their Teachings.—Because the directions of weakness are geometrically fixed for each crystalline system, and because those of each system differ from those of the others, the silhouettes vary correspondingly with the system, and hence may be used for determining the system to which a substance belongs. Because for given crystalline system the directions of weakness vary with the direction which the surface on which the depression is made, that is the polished surface of the specimen, bears to the crystalline axes of the substance, the silhouettes vary correspondingly. Because the direction of the crystalline axes, that is the orientation, varies from polygonal grain to grain in a given polished metallic section, the silhouettes vary in like manner from grain to grain.

But if a mass can be had composed of a single grain and hence oriented alike throughout, and large enough for making and examining the silhouettes made on two or three plane faces which have different directions, then both their shape and the relation which their shapes on these various faces bear to each other may be made to disclose not only the crystalline system to which the grain belongs, but the direction which each face bears to the

crystalline axes of that system. Again, knowing the direction which any given face bears to the crystalline axes, for instance, that it is parallel to two of them and hence is a cubic face, and knowing the direction of the intersections made on a cubic plane by the planes of every other crystalline form, we can, by determining the direction which any given line follows on that face, infer what the form is to the faces of which the plane tracing that line is parallel, for instance, whether that line is the intersection of an octahedral plane with the surface of the specimen.

435. The Silhouettes in Iron.—Proceeding thus, Osmond and Cartaud[1] succeeded in showing that the silhouettes of alpha iron or ferrite, gamma iron or austenite, and also of the metal in the alleged beta range of temperature, are those of the isometric system, as shown in Figs. 38, 39 and 40 (p. 270) which represent these three states. The indentation was done on ferrite at about −185°, at room temperature, at about 300°, and at 600°; on beta iron at about 800°; and on austenite at 900°, and also in the cold on manganese and nickel steels. In each of these three figures the face A or $tt'rq'q$ is part of a cubic face, of which tt', $t'r$, and qt are the whole or part of the sides. Face B or $qq'q''$ is the octahedral truncation of a cube (§837). That is to say the front lower left-hand corner of this cube was cut off along an octahedral plane, and on this truncation a silhouette was formed by pressure. In order to illustrate this silhouette simultaneously with that of the cube face, the octahedral plane is supposed to be developed by swinging it about the line qq' into the plane of the paper.

Face C, or $q'rr'r''$, in like manner is a rhombo-dodecahedral truncation of a cube (§840). That is to say, the lower right-hand edge of the cube is sliced off along a rhombo-dodecahedral plane, and on the truncation thus formed a silhouette is made. This is shown by swinging the truncation about the line $q'r$ into the plane of the paper.

In each case the silhouette was made by pressure on a sharp needle point, ground to an angle of 60°.

436. On the cubic face, A, the silhouette is a St. Andrew's cross, with four axes of symmetry,[2] two parallel to the diagonals of the square of which the face of a perfect cube would consist, and two parallel to the sides of that square.

The arms of the cross, and those of its elements which consist of straight lines, are parallel to the diagonals of the square.

[1] *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, p. 444.

[2] By an axis of symmetry is meant a line about which the silhouette is symmetrical. For instance, if you fold the paper along the line $t't''$ of Fig. 38 and close the fold by bringing the two parts of the paper together and pressing the paper at the right hand of the fold down upon the paper at its left, the lines of the right-hand half of this figure will fit down upon those of the left-hand half and match them. Therefore $t't''$ is an axis of symmetry of face A. The same is true not only of the line which forms the other diagonal of the square, but also of a vertical and of a horizontal line drawn through the center of the figure. On the face C there are only two axes of symmetry, the lines uu' and vv'. That half of the figure on the right of each of these would fit and match the half on the left side if folded down upon it. But this is not true of any other line which can be drawn across the figure.

437. On the octahedral truncation, *B*, of this cube the silhouettes are tripartite, with three axes, each bisecting one of the angles. In both ferrite and austenite the best marked straight elements of the silhouettes are parallel to the sides of the triangle.

438. On the rhombo-dodecahedral truncation of a cube, face *C*, the silhouette is like an hour-glass, and has two axes of symmetry, *uu'* and *vv'*.

Thus the silhouette of each of these three crystallographic faces has the same axes of symmetry and the same general shape in all three states. There are indeed important differences. For instance the straight elements of each of the three crystallographic sections are much less pronounced in beta than in alpha and gamma iron, and a like difference exists between the alpha and the gamma rhombo-dodecahedral truncations. Differences of detail occur in each of these nine sections, that between the upper and lower halves of *C* of Fig. 40 being very marked.

439. The Elements of the Silhouettes are Slip Bands.—Each silhouette is made up of a great number of very fine lines which, in the ferrite of Fig. 40, must be slip bands and not Neumann bands, because they were caused by quiescent pressure which cannot cause Neumann bands in this specific ferrite, as Osmond and Cartaud assert explicitly.

440. The Silhouettes and the Slip Bands are Geometrical.—The important thing is first that these silhouettes have these axes of symmetry; second, that the relation between these three faces as regard these axes is that which holds true of the isometric system; and third, that on each of three faces cut at right angles to each other the silhouettes have four axes of symmetry,[1] which is possible only in the isometric system. These facts are cogent evidence that the silhouettes are geometrical, and hence that the straight elements of those slip bands which compose them also are geometrical. It does not follow that all slip bands formed under all conditions are geometrical.

441. Additional Evidence that the Slip Bands are Geometrical.—The three sets of slip bands in the crystal of native copper shown in Fig. E of Plate 45 form the three sides of an equilateral triangle, as the intersections

[1] Osmond and Cartaud, *idem*, p. 457. There was formerly a belief that gamma iron crystallizes in the rhombohedral (hexagonal) system, in rhombohedra simulating cubes, as bismuth does. But this belief had to be set aside when Osmond and Cartaud thus found four axes of symmetry on each of three faces of gamma iron cut at right angles to each other, because this cannot occur in any crystal of the hexagonal system, or for that matter of the orthorhombic (trimetric) system. It is true that while these investigators assert categorically the existence of quadripartite symmetry on two faces cut at right angles, they do not so assert its existence on three such faces, but leave it to be inferred with confidence from their language, which indeed often avoids the categorical form.

The reason why gamma iron was suspected not to be isometric is that allotropic modifications corresponding to higher temperatures very often belong to some system of a lower order of symmetry than that of the low temperature state, the isometric system having the highest order of symmetry, in the sense that a given crystal is symmetrical about a greater number of planes in this system than in any other.

A second reason assigned for this suspicion is that gamma iron is very subject to twinning. But this reason seems without weight, because some of our most familiar isometric native minerals very often twin, for instance spinel, tetrahedrite, fluorspar, and galena.

of the octahedral planes by three other octahedral planes or by three cubic planes do (see Fig. 46). In idiomorphic crystals within a lead-bismuth alloy Ewing and Rosenhain developed by deformation geometrically arranged lines, bearing a definite relation to the axes of the crystals.[1] They further inferred that the slip bands were parallel to the octahedral cleavages as they are habitually in ferrite, and as Mügge finds that they are in crystals of native copper.[2]

Further, Ewing and Rosenhain found slip bands in parallel series extending across apparently disconnected patches of the white constituent of the lead-bismuth eutectic, disconnected, that is to say, in the plane of actual microsection,[3] though probably connected outside of that plane. This parallelism accords with the uniformity of orientation of such apparently disconnected patches of a constituent of a eutectic, but it is not direct evidence of the geometrical arrangement either of this constituent or of the slip bands, because ungeometrical lines also may be parallel.

So, too, the occurrence of slip bands in straight and parallel lines, and the fact that these lines show no clear preference for parallelism to the grain boundaries,[4] are wholly in conformity with the belief that the arrangement of the slip bands is geometrical, but not direct proof that it is.

Humfrey,[5] operating on very large and well-oriented dendrites of lead, found that the slip planes were perpendicular to the octahedral axes of the cubo-octahedral etching pits, the cubic faces of which in turn were parallel to the main axes of the dendrite. Here the evidence is cogent. The markings seem clearly to have been slip bands and not twins, but in the case of the lead-bismuth alloy of Ewing and Rosenhain just noted this is not clear.

442. To What Crystallographic Directions are the Slip Planes in Iron Parallel?—This question may be answered by a study of these silhouettes, Figs. 38 to 40. To this end I assemble beside them eight figures, numbers 43 to 50, showing the direction followed on a cubic face and on octahedral and rhombo-dodecahedral truncations of a cube, that is on truncations like *A*, *B*, and *C* of Figs. 38 to 40, by the intersections of those truncations by cubic, octahedral, and 211 planes. The derivation of certain of these intersections is indicated in Appendix I.

This assembling makes it easy to compare the directions actually followed by the elements of these silhouettes with the theoretical directions which the intersections of these important planes ought to follow on such truncations, and thus to infer to what crystalline form the planes belong of which

[1] *Phil. Trans. Royal Soc.*, 1900, A, vol. 193, p. 372, and Figs. 50–51; *The Metallographist*, 1900,vol. 3, p. 128, Figs. 43 and 44.

[2] Martens-Heyn, "Materialenkunde," vol. II A, Springer, Berlin, 1912, p. 223.

[3] *Op. cit.*, p. 371 and Fig. 49; p. 127 and Fig. 42.

[4] Rosenhain, *Journ. Iron and Steel Inst.*, 1904, No. I, vol. 65, p. 349. It is true that Osmond, Fremont, and Cartaud believed that, under greater deformation, additional and like lines appeared in different directions, but nearly always parallel to one of the boundaries of the grain or of a neighboring one (*Rev. de Metallurgie, Mem.*, 1904, vol. 1, p. 27).

[5] *Phil. Trans. Royal Soc.*, 1903, A, vol. 200, p. 225.

these slip bands are the outcrops or intersections with the known planes represented by A, B, and C. In order to make this still easier, the sides of the figures showing these theoretical intersections are set so as to be parallel to the corresponding parts of Figs. 38 to 40; that is to say the cube is supposed to stand on one face, so that two of the sides of the cube face shown are vertical and two horizontal, and the octahedral and rhombo-dodecahedral faces are inclined as they are in Figs. 38 to 40, and hence are supposed to be derived from that same cube by truncating its front left-hand lower corner with an octahedral plane and its right-hand lower edge with a rhombo-dodecahedral plane, and swinging the sections thus cut into the plane of the paper. To carry out this idea, the corners of these several figures are lettered to correspond to Figs. 38 to 40, and parts of two sides of the cubic face are sketched in each figure.

Note that qq' and $q'r$ are intersections of the cubic face $tt'rq'q$ with an octahedral and a rhombo-dodecahedral face respectively. This serves to fix one direction on each of these planes.

443. The slip planes in austenite are octahedral, according to Osmond and Cartaud.[1] Their silhouettes, Fig. 38, bear out this inference fully. On the face, A, the slip bands, that is the straight elements of the silhouettes, are diagonal only. This is not decisive, because, though these diagonals are not the intersections of cubic planes (Fig. 43), they may be of either octahedral or 211 planes, as shown by Figs. 43 and 44.

On the octahedral truncation, B, each slip band is parallel to one or another of the three sides. Hence their direction teaches us nothing, because while this direction is true of the octahedral intersections of an octahedral truncation (Fig. 46), it is true also of all the cubic intersections of such a truncation (Fig. 46), and of three of the 211 intersections, numbered 1, 4, and 7 in Fig. 47.

On the rhombo-dodecahedral truncation, C, the indications are far more decisive. Thus the three directions which the slip bands follow are parallel or nearly parallel to the three directions, 1, 2, 3, of the octahedral intersections of such a face, as shown in Fig. 49. This is shown more clearly by plotting on face C of Fig. 38 these three octahedral lines 1, 2, and 3 from Fig. 49. The slip bands parallel line 1 exactly, and they are nearly parallel to lines 2 and 3. Thus all the octahedral intersections of such a truncation are represented by these slip bands, and every slip band is parallel to one or another of those octahedral intersections. The correspondence is complete quantitatively, though not qualitatively because the parallelism is not exact between the oblique slip bands and lines 2 and 3. The fact that one set of the slip bands is parallel to a cubic intersection of this truncation, $q'r$, or line 1 of Fig. 48, is of course unindicative, because this direction is at the same time octahedral. Had the slip bands been cubic, the second cubic intersection, 2 of Fig. 48, might have been expected.

[1] *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, p. 468.

So with the fact that one set is parallel to line 1 of the 211 intersections of such a truncation, Fig. 50.

You can assure yourself that the oblique lines of C are not 211 lines by drawing on tissue paper the 211 lines of Fig. 50, together with its sides rr', $r'r''$, $r''q'$, and $q'r$, and superposing this tracing on C. You will then see that the 211 lines, 2, 3, 4, and 5, most nearly parallel to the slip bands in C, do not match the slip bands so well as the octahedral lines 1, 2, and 3 do.

To sum up, all the intersections of octahedral planes on cubic planes and on octahedral and rhombo-dodecahedral truncations of a cube are represented by the slip bands actually present in austenite, in Fig. 38, and all these slip bands actually present in that austenite correspond to such octahedral intersections, further corresponding only to such cubic and 211 intersections as coincide in direction with these octahedral ones. Hence the slip bands are taken to be intersections by octahedral planes.

Additional evidence that the slip bands in austenite are octahedral is found in the observation of Osmond and Cartaud that when slip bands and twins coexist here, one set of slip bands is always parallel to the twins, which in austenite are known to be octahedral.[1] If these writers here refer to random sections, as is almost certain from their sketch, this implies that at least one set of the slip bands is octahedral, because though, as we have seen, the octahedral intersections may, on certain definite crystallographic planes, be parallel to those of other planes, being for instance parallel to the cubic intersections on an octahedral truncation, as in Fig. 46, and to one of the cubic intersections on a rhombo-dodecahedral face, yet on random planes they could only rarely and by accident be parallel to the intersections of other crystallographic planes.

444. Other Manifestations of the Octahedral Structure.—Because at all temperatures above $A3$ and SK of Fig. 23, that is at all temperatures high enough to cause marked mobility and hence easy migration, the mass is austenite, except such free cementite as may be present, this octahedral or Widmanstättian structure is extremely prominent under various conditions. This is the pattern followed by the pro-eutectoid ferrite for instance in Fig. A of Plate 28; by the pro-eutectoid cementite in Figs. G to I of Plate 3; and by the martensitic markings of Figs. A of Plate 2 and F of Plate 28. For like reasons the structure of alloys which are austenitic at the room temperature, such as 25 per cent. nickel steel, invar, and quenched manganese steel, is octahedral, and hence their twins, both annealing and mechanical (§§639 and 652), their slip bands as we have just seen, and the assembling of precipitated cementite (§655) are octahedral.

445. The Slip Bands in Ferrite Follow the 211 Trapezohedral Planes.—Ewing and Rosenhain[2] agree with Osmond and Cartaud in finding octahedral directions for the slip bands in ferrite, but the former observers differ from the latter in finding also cubic ones. That is to say, they find that though

[1] *Idem*, p. 468.

[2] *Phil. Trans. Royal Soc.*, 1900, A, vol. 193, p. 365, and Figs. 29 and 30.

after slight deformation the slip bands are rarely parallel to the sides of the (cubic?) etching pits and hence rarely cubic, yet after severe deformation one set of slip bands is thus parallel to those sides and hence cubic, while (two?) other sets are diagonal to them and hence not cubic but octahedral. Those parallel to the pits could not be octahedral but might be cubic, while those diagonal to them could not be cubic but might be octahedral. Hence their inference that the slip planes are octahedral, and also though more rarely cubic.

These observations carry less weight than those of Osmond and Cartaud, both because they were apparently made on random sections and because of the absence of supporting micrographs and sketches.

These latter observers[1] seem to base their findings on their silhouettes, such as those of Fig. 40. But these, I believe, fit the theory that these slip bands follow the planes of the {211} trapezohedron. Let us now consider the elements of these silhouettes in detail.

446. On the cubic face, *A*, of Fig. 40 the slip bands are exclusively diagonal, and hence cannot be cubic, for in that case they would be parallel to the sides, as in Fig. 43, but they might be either octahedral or 211, as shown in Figs. 43 and 44.

447. On the octahedral truncation, *B*, the best marked slip band directions, 1, 4 and 7, are parallel to the sides of the truncation, and therefore might be either cubic, octahedral, or 211, as shown in Figs. 46 and 47. Note in this last that the lines numbered 1, 4 and 7 are thus parallel to the sides of the truncation. Moreover, there is on the left-hand side of the silhouette of *B* a pair of lines, 3 and 5, none too straight it must be admitted, yet well marked, and corresponding to a like pair on each of its other sides. 3 and 5 of *B* are apparently parallel to 3 and 5 of Fig. 47. They make with each other an angle much like the 141° 48′ angle which that other 3 and 5 make, and they make with the sides qq', $q'q''$ and $q''q$ and with lines 7 and 1 of *B* an angle much like the 79° 6′ angle which that other 3 and 5 make with the corresponding lines of Fig. 47. In like manner on the right-hand side of the silhouette there are lines parallel to lines 2 and 9 of Fig. 47, and on its lower side there are others parallel to lines 6 and 8 of that figure. Hence these lines in the silhouette correspond to those in Fig. 47 in the angles which they make with each other and with the sides of the octahedral truncation.

This is seen more clearly by reproducing in Fig. 41 these three pairs of lines from the silhouettes, together with the octahedral directions, shown in dotted lines, and two different groupings of the 211 lines, here drawn unbroken and numbered as in Fig. 47. The angle between the two lines of each pair from the silhouette is very much flatter than the octahedral angle of 60°. Moreover, no possible combination of octahedral lines can give any angle other than 60°, unless perhaps through the coexistence of two sets of octahedra by twinning, so that an octahedral interpretation is difficult.

[1] *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, p. 461.

Those silhouette angles are also flatter than the 3–8, 9–5, and 6–2 groupings of the 211 lines. But if we exclude the very intersection of the silhouette pairs and consider the major part of their length, we may admit a fair correspondence between them and the flatter of the 211 groupings, 2–9, 8–6, and 5–3.

If we admit this, then all nine of the 211 intersections of an octahedral truncation are represented among these slip bands, and there is no straight part of any of these slip bands which does not correspond to some one of these 211 intersections. Thus all the facts fit the theory that these slip bands are 211. Though three of these slip band directions correspond also to the cubic and octahedral intersections of such a face, these are the intersections common to the cubic, octahedral, and 211 planes, and hence are wholly consistent with the theory that these slip bands are 211.

448. On the rhombo-dodecahedral truncation, *C*, of Fig. 40 the nearest approach to straight elements are the sides of the re-entering angles *c*,*c*′ and *e*. The cubic intersections, 1 and 2, of such a truncation are parallel to its sides, as shown in Fig. 48. This line 2 has no corresponding line in the silhouette, and line 1 is wholly unindicative for our present purpose, because if any corresponding line can be traced in the silhouette it corresponds equally to this cubic line, to line 1 of the octahedral intersection of this present rhombo-dodecahedral truncation, and to line 1 of the 211 intersection of this truncation. Therefore we may leave line 1 out of consideration.

Finding thus no evidence that the slip bands on this truncation are cubic, let us see how they correspond to the octahedral and to the 211 intersections of a rhombo-dodecahedral truncation, omitting this line 1 of each. Of the octahedral intersections of this truncation there remain two, numbered 2 and 3 in Fig. 49, and of the 211 intersections there remain four, numbered 2, 3, 4, and 5 in Fig. 50.

In order that we may be able to compare the silhouettes with these six lines, I draw in *C* these two octahedral lines of Fig. 49 with dashes at 2 and 3, and these four 211 lines of Fig. 50 in solid lines at 2 and 5, and also at 3 and 4. Note that each of these lines as drawn on *C* is parallel to the line of the same number in Figs. 49 or 50 as the case may be.

The straight elements which make up the angles *c* and *c*′ in *C* seem to me fairly parallel to the 211 lines 2 and 5, and thus to correspond to these 211 lines much more closely than they correspond to the octahedral lines 2 and 3. So, too, the straight elements which make the slip band angle *e* are fairly parallel to the 211 lines 3 and 4, and thus correspond to these 211 lines, but they are not parallel to any of the octahedral lines, forming an acuter angle than that between the octahedral lines which they most nearly resemble, 2 and 3. This correspondence is clearer in the original.

To sum this up, the line 1, common to the cubic, octahedral, and 211 intersections of this rhombo-dodecahedral truncation, *C*, being set apart because of its unindicativeness, no other of either the cubic or the octahedral

intersections of such a truncation is represented among the slip bands actually present, yet all four of the other 211 intersections of such a truncation are represented among these slip bands, and each of the straight parts of these slip bands corresponds to one or another of these four 211 intersections. Thus there is no suggestion here that these slip bands are either cubic or octahedral, but every evidence that they are 211. Quantitatively the evidence is all one way, though qualitatively it is not of the best, because of the curving of these elements in face *C*.

449. The Three Planes of Ferrite Collectively.—To sum up the foregoing, first the slip bands on all three planes agree in having no direction which is either cubic or octahedral except those which are also 211 and hence unindicative; and second, those on truncations *B* and *C* agree not only in having directions which correspond to all the 211 intersections of such truncations, but also in that each of their various directions corresponds to one or another of such intersections.

A indeed lacks the lines running from the corner to the middle of the opposite sides of the square, the 2:1 lines, shown in Fig. 44. Yet these same lines are often lacking, for instance, among the Neumann bands, though these are certainly 211. Indeed the angle, 60°, of the needle employed by Osmond and Cartaud in developing these slip bands was such that it would be less likely to develop these directions than the diagonal ones, as is seen on referring to Fig. 119, p. 595. On the upper face of this cube a diagonal line would be caused by slip along a plane making an angle of 35° 16′ with this surface, whereas a 2:1 line would be caused by slip along a plane making an angle of 65° 54′ with this surface. The sides of the needle-point making an angle of only 30° with the vertical would evidently set up a much greater resultant on those flatter planes which make the diagonal intersections than on those steeper planes which make the 2:1 intersections. Hence the presence of the diagonal and the absence of the 2:1 intersections.

Here, I think, are rather strong indications that the slip bands in ferrite resemble the Neumann bands in usually representing not cubic or octahedral but 211 planes. The observations on which Osmond and Cartaud seem to have relied really point strongly in this direction.

The directions which Ewing and Rosenhain took to be octahedral, diagonal to a cubic pit and hence to a cubic face, are unindicative here for they correspond exactly as well to the 211 as to the octahedral direction. Their occasional cubic directions are not improbable, in view of my finding that the habitually 211 Neumann bands are occasionally cubic (§591).

Hence the present evidence indicates that these slip bands habitually follow the planes of the 211 trapezohedron, and not the octahedral planes as has hitherto been reported; yet the occasional cubic directions reported by Ewing and Rosenhain, though as yet unconfirmed, are not improbable.

This accords with the facts that the Neumann lamellæ follow these same planes, and that the fracture, though habitually following cubic

planes as in Fig. G of Plate 15, sometimes follows 211 planes, as in Fig. D, Plate 32. Note that the end of the prism *A* of that figure is bounded by a line which, though it looks in the photograph like the diagonal of a square, is in fact approximately a 2 : 1 line, like those of Fig. 44.

450. The slip bands in beta iron, Fig. 39, are curved so smoothly and regularly as to give no important indication as to their crystallographic direction. The lines most nearly straight are those following the diagonals of the cube face, which might be either octahedral or of the 211 trapezohedron, but not cubic. Their being curved is not due to temperature as such, because practically pure iron above 900° yielded straight deformation lines, Fig. 38, face *C*.

451. Ungeometrical Slip Bands.—Though the straight parts of the slip bands in Figs. 38 and 40 are thus evidently geometrical, there are many slip bands which may not be. Apart from their often being strongly and irregularly curved, their general direction may be difficult to connect with any probable crystallographic direction. Thus the silhouette of ferrite in Fig. E of Plate 26, much like that of face *A* of Fig. 40 in its appearance and apparently in its preparation, has straight elements which the reasoning in §§445 to 449 leads us to refer to the planes of the {211} trapezohedron. After this silhouette had been developed, the specimen underwent a new but slight plastic deformation, setting up new slip bands cutting across the field, including the cross itself, but parallel neither to the sides of the square nor to its diagonals, that is to the axes of the cross arms, nor yet to any line from a corner to the middle of an opposite side, and hence following neither cubic, octahedral, nor 211 trapezohedral planes,[1] those which might be expected to be represented in the crystalline organization of this ferrite. Note that these slip bands are far from parallel to those of the 211 lines which differ least from them, those at AA' and BB'. Moreover, they are much more crooked than the straight elements of the cross in this same figure.

The suggestion here is that we have to do with two distinct things, slip along the crystalline planes, yielding the straight elements of the cross, and some other mechanism of deformation, yielding the crooked bands apparently following random directions. A like interpretation then suggests itself for the silhouettes in Figs. 38 and 40, that the straight elements following definite crystallographic planes represent true slip, but that the curved ones do not. This will be considered further in §§463, 498, *et seq.*

452. Ungeometrical Silhouettes.—Closely related to these ungeometrical slip bands are the ungeometrical silhouettes which Stead developed.[2] He cut a bar from a crystal of iron, twisted it about its axis, and cut and polished a longitudinal section parallel to that axis, and then pricked it to develop the silhouettes. Had the crystal units rotated about their axes during their rotation about the axis of the bar, or in other words had their motion

[1] Compare Figs. 43 and 44.

[2] Private communication, Dec. 5, 1914.

been unvectorial, this would have made the silhouettes very irregular, whereas had the motion been strictly vectorial, each crystal unit continuing to face exactly in its initial direction, the regularity of the silhouettes need not have been lessened by the twisting. In fact they varied greatly from a regular cross like that of face *A* of Fig. 38 to figures little related to it. Here is a strong suggestion of unvectorial or fluid motion.

CHAPTER 18

SLIP BANDS, DISCUSSION

453. Introduction.—After considering in §§454 to 461 the mechanism of crystalline slip, I attempt in §§462 to 477 to answer the question which till now I have left open, whether what are called "slip bands" really result from slip.

454. Slip a Possible Mechanism of Plastic Deformation.—It is currently accepted that plastic deformation of non-metallic crystalline bodies, in addition to taking place by the usual "twinning" mechanism explained in Chapter 24, may occur through crystalline slip. Thus a bar cut from a crystal of ice with its axis parallel to that of the crystal, and loaded as shown in Fig. 62, sags slowly into the position there shown in broken lines. But in the bent part the crystalline orientation is still parallel to the initial, so that the bending is evidently an accumulation or integration of a great number of minute slips along the slipping planes normal to the axis, as indicated in Fig. 62*C*. Many minute crystalline blocks forming part of the initial crystal have slipped past each other without losing either continuity or their initial orientation. How continuity is retained during this slip will be explained in §508.

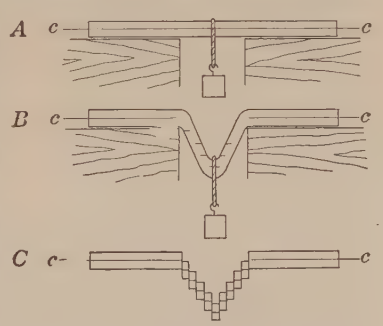


FIG. 62.—*A* and *B*. The intercrystalline sagging of ice under a transverse load. *C*. This sagging is by intra-crystalline slip along the cleavage planes. After Linck. (From Bayley, *Elementary Crystallography*, McGraw-Hill, 1913, p. 183).

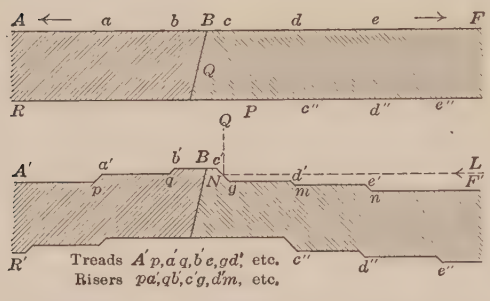
NOTE TO FIG. 62.—*cc* is a rod of ice with its axis parallel to the vertical axis of the crystal of ice from which it is cut. In the bent part the longitudinal axis, shown by the horizontal broken lines, retains its initial direction.

In the same way the "liquid crystals" of para-azoxyanisol, like those shown in Fig. B of Plate 14, when deformed by their own surface tension into spheroidal drops, remain doubly refracting, whence we may infer that their internal orientation either has persisted undisturbed through this deformation or else has re-established itself after the deformation, perhaps propagating from the undisoriented interior parts of the drop to its outer parts which have been deformed. The optical methods by which slip without rotation can be distinguished in these transparent substances from twinning are evidently inapplicable in their present form to metals.

455. The Supposed Mechanism of Crystalline Slip in Metals. Treads and Risers.—For simplicity we may assume that Fig. 63 represents part of the longitudinal section of a thin sheet of metal which is pulled as shown by the arrows, and that in this part it is composed of two grains *AB* and

BF, each of which extends quite across from one side of the sheet to the other.[1]

The structural planes along which slip can occur, the "slip planes," in grain *AB* point down and to the left, those in grain *BF* point down and to the right. Ewing and Rosenhain conceive that, on passing the local elastic limit, slip occurs along certain of these planes as shown in Fig. 64, creating the little steps *a'*, *b'*, *c'*, *d'*, and *e'* with their *treads* *gd'*, *me'*, *nF'*, representing fractions of the initial polished surface, and their *risers* *c'g*, *d'm*, and *e'n* representing parts of these "*slip planes*" or planes along which the slip has occurred. The risers, because they are inclined and hence cannot reflect the vertical light vertically back into the microscope, appear as narrow black bands. On this conception these risers are the slip bands.



FIGS. 63 AND 64.—Idealized intra-crystalline slip in a coarse-grained steel tensile test-bar.

When a specimen on which slip bands have been developed is rotated on the stage of the microscope while under oblique illumination, the various systems of slip bands flash forth successively like revolving light-house lights. When this rotation reaches a certain stage bright lines occur where the black ones were originally.

These dark lines clearly are not cracks, first because they are not effaced by annealing, though this brings the metal back almost exactly to its initial strength and other physical properties, and second because this overstrain or incipient plastic deformation actually strengthens the metal greatly, which it would hardly do if it caused internal cracks.

Some of the incredulous have thought that the slip bands might be etching lines along the edges of twins caused by the deformation. But first the slip bands appear on the polished surface without etching; and second, etching very often fails to detect traces of slip, though it would if the slip occurred through twinning.

456. Similes to Explain the Mechanism of Slip.—We may conceive the cement joints in a thick brickwork mass replaced by wax guides with limited shearing strength, so that under strong pressure any row of bricks can slip past any other and thereby shear across a series of wax guides normal to this movement, while it is yet held as by irresistible magnetization so that it can neither rotate nor deflect. The guide can be sheared across; it can move with the bricks in any direction as long as it remains parallel to its

[1] Steel in which the grains are so coarse as to reach thus from face to face of the specimen is indeed usually so extremely brittle that it breaks with very slight permanent deformation. Because its capacity for this deformation is so slight, the degree to which slip bands, the product of that deformation, can be developed in it is correspondingly slight. Such material, therefore, though it fits the purposes of description, is ill fitted for the actual study of slip.

initial direction; but from that direction neither the guide nor the bricks which it guides can turn.

The passing of the elastic limit means that the stress reaches such intensity that certain crystalline slices, forming part of certain crystalline grains, start to slide along the slip planes over the similar neighboring slices in those same grains.

There is of course no wax guide between the sliding blocks, but they slide as if there were such a guide.

Again, the tapering punch A, Fig. 65, pushes down against the face st of a cube 3 at the upper surface of the metallic mass. It pushes that top cube, 3, and each of these below it, $3A$ to $3C$, straight down, so that their bounding planes M, E, F, G, and H come to occupy the positions shown in broken lines at M', E', F', G', and H', the punch thereby reaching the position shown in the inclined broken lines.

Because each of these little cubes is microscopic, we must next substitute in our minds a needle point for the tapered punch A.

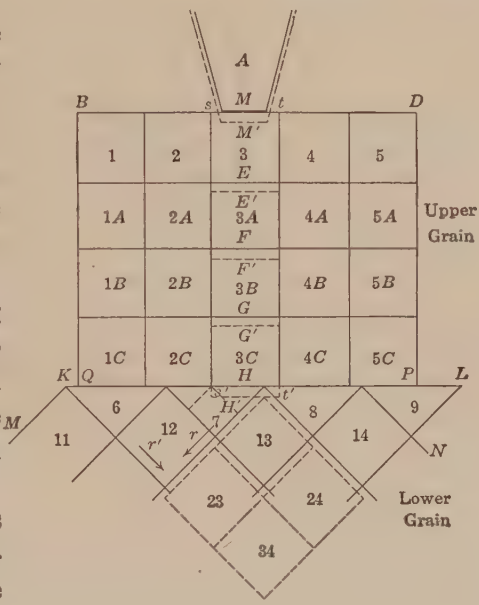


FIG. 65.—Slip of a row of crystalline blocks normal to the surface.

Let us next suppose that A represents not the sharp point of a needle but the cross-section of a knife blade, the length of which is normal to the plane of the paper. Then not only row 3 but a whole slice, comprising a series of rows just like row 3, but lying behind the plane of the paper, would in like manner be forced down, as a card might be forced down through a pack standing on edge, or a book drawn from a full shelf. The contact planes between this book and those on either side of it are in like manner slip planes.

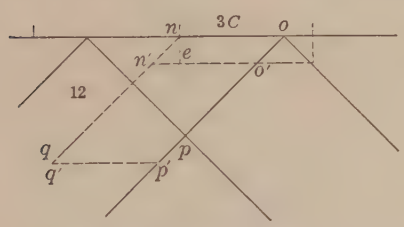
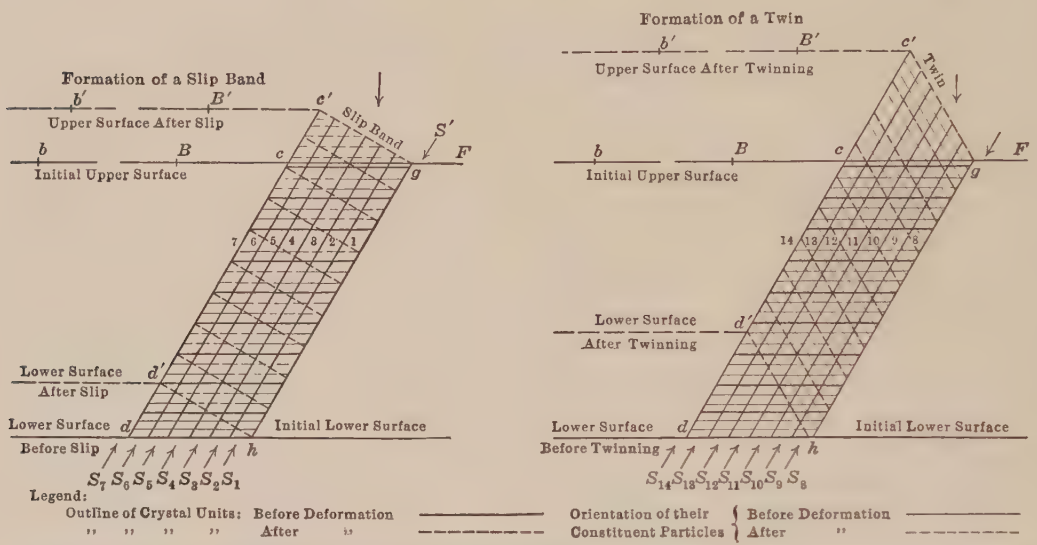


FIG. 66.—Detail of Fig. 65.

Moreover, the downward movement is not confined to the vertical row of blocks, 3 to $3C$, against which the imaginary narrow edge A presses, but spreads to the adjoining ones, which are dragged down along with this axial row, though each successive vertical row descends less than its neighbor nearer the axis of the punch, quite as pulling your gloved hand out of your pocket tends to pull the pocket along with it. And with this downward movement of these rows come complex movements through the mass, partly centrifugal in the regions which the punch is approaching, partly centripetal in those which it has lately passed. This centripetal pressure forces the sides of the punched hole so firmly against the punch that they are bur-

nished by its further descent, while the vertical component of the complex motion roughens a circular figure on the upper surface, *BD* of Fig. 65, in the way shown in Fig. E of Plate 21.

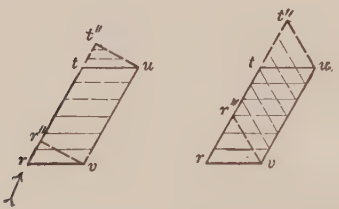
457. An Alternative Mechanism of Slip after Osmond and Cartaud.— Fig. 67 is supposed to represent a detail of step *d'm* of Fig. 63, and Figs. 69



Figs. 67 and 68.—Mechanism of the formation of slip bands and twins as inferred from the verbal description of Osmond and Cartaud.

to 71 in turn to represent a single crystal unit, for instance of slice 1 of Fig. 67.

The initial shape of a unit as shown in Fig. 69 is *rtuv*, parallel to all the other units in Fig. 67 in their initial position as sketched at the lower ends of slices 1 to 7. The side *uv* of Fig. 69 remaining fixed, stress in the direc-



Figs. 69 and 70.—Slip and mechanical twinning. Detail of Figs. 67 and 68.

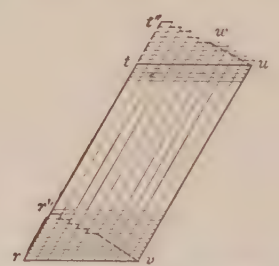


Fig. 71.—Inequilibrium of the constituent particles in slip-band-forming crystal unit.

tion of the arrow slides the side *rt* in its own track upward and to the right, thus twisting the unit into the new shape *r''t''uv*. This twisting of course carries with it the various constituent particles of the unit, but these, instead of rotating each about its center and so changing orientation, move only vectorially, changing their position in space but not the direction in which they face. Hence after the slip has occurred the unit has the outward

shape $r''t''uv$ and the internal orientation represented by the unbroken lines, an orientation which is clearly out of equilibrium.

Two conceptions of this state of inequilibrium are shown in Fig. 71, in which as before the broad lines show the form of the crystal unit and the faint ones the form of its constituent particles. The ends of those particles may form a series of steps as from t'' to w; or the end particles, acting as if they were of india rubber, may have their upper surfaces smoothed by surface tension into one continuous straight surface as sketched at wu, without any change in their volume.

Turning now from the behavior of a given unit to that of the various crystallographic slices, 1 to 7, which collectively make up the region involved in the formation of any given slip band, as in Fig. 67, there is a twisting of the outer form of the several units which make up the various slices, brought about by stresses $S1$ to $S7$, which changes the outer form of the aggregate mass involved, from $cghd$ to $c'ghd'$. Yet the constituent particles of these units retain their initial orientation indicated by unbroken narrow lines at the bottom of the slices, so that the condition after twisting is that shown at the top of these slices, a twisting of the outer form of the various units from their original shown in unbroken wide lines to that shown in broken wide ones, but with the retention of the initial orientation of their constituent particles as shown. Thus the initial state of equilibrium in which the internal orientation of each unit corresponded to its outward form is replaced by one of inequilibrium, in which the internal orientation does not fit the outer form.

458. Discussion of the Slip Band Theory of Osmond, and Cartaud.[1]— For the simple slip conceived by Ewing and Rosenhain, as of half a pack of cards over the other along the plane $c'g$ of Fig. 67, this present theory substitutes what is in effect incomplete twinning, a rocking or rotating of the units which compose each of the slices of metal involved in the movement, each unit about its right-hand side, together with a lifting of each unit by its own rocking and by that of those at its right. Here the part cg of the surface of the specimen wheels to the right into the position $c'g$. In this view there is no slip plane, but a succession of gliding planes from gh to cd. The direction of the motion is approximately normal to the riser $c'g$, instead of being parallel to it as in simple slip. This may be called the "wheeling" as distinguished from Ewing and Rosenhain's "slip" theory of slip bands.

The merit of this theory as regards its binding slip and twinning together as different stages of a single process of deformation can be discussed better in Chapter 24, after we have considered twinning. I find it less easy than the slip theory to reconcile with the failure of slip bands to reappear on repolish-

[1] *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, p. 447. My description represents my understanding of their slip band theory, after filling up as best I can the gaps in their very condensed enunciation of it. The sum of the 7 slices which thus slide they call a "mesh." They say that it changes its form by "simple translation," but that the "molecular polyhedra," which we may take as identical with the crystal units, "retain their initial orientation after displacement."

ing and etching. On the slip theory, the region disturbed along the slip plane may be only some molecules in thickness, and hence so thin that any difference in etching tint between it and the enclosing metal could not be detected on any possible magnification. But on this wheeling theory the thickness of the involved region is of a far higher order, for instance that of the width of the risers of Fig. E of Plate 25. Because the metal in the involved region is in inequilibrium according to this theory, it might well be expected to etch to a tint different from that of the enclosing metal, and this difference should be readily visible because of the very appreciable thickness implied for the wheeling region. Thus, finding the generally accepted slip theory on the whole better in accord with the facts than the wheeling theory, I adopt it in what follows.

459. The Movement not of Crystal Units but of Blocks of Units.—Having in mind this conception of slip as sketched in Fig. 65, let us next recognize that the crystalline mass of each of the polyhedral grains, shown in Fig. *F* of Plate 14, of which the metal consists, is made up of great numbers of crystalline units, meaning by a unit the smallest mass capable of crystalline arrangement. Such a unit may on one hand be far greater than a molecule, or on the other hand it may be only of an atomic order of magnitude, a conception which Bragg's researches support. In the movement which we have considered, certain relatively large blocks of these crystalline units move each as a whole and without changing the relative positions of their units, as if only certain of the cement joints of our masonry of §456 were thus replaced by these extraordinary wax guides, while between these guides were still many, perhaps hundreds or thousands of cement joints, not thus replaced, yet capable of such replacement in case pressure along them became sufficiently great, or as half the leaves of a book bound weakly in paper can be made to slide past the other half, if we arrange the shearing stress so that it overcomes the resistance of the weak binding.

460. Slip in Grains with Orientation Oblique to the Stress.—Up to this point our study has been simplified by assuming that in the grain attacked the orientation is parallel to the stress, so that the cubes of row 3 of Fig. 65 have one set of their faces parallel to the upper surface *BD*, and two other sets normal to it, and therefore serving as natural surfaces of slip. Below this grain lies another of which the upper part *KLNM* is sketched, with orientation inclined at say 45° to that of the path of the punch or knife, and to the orientation of the upper block *BDPQ*. The descent of block 3*C* forces down the triangular blocks 7 and 8 and cube 13, and the descent of these in turn forces down cubes 12, 14, 23, 24, and 34. Thus a general readjustment occurs throughout a wide region, radiating from the direct path of row 3.

Here again not only the cubes of row 3, but also the various cubes of the lower grain *KLNM* in moving down refuse to rotate on this theory of crystalline slip. For instance, cube 12 accomplishes its descent by sliding over its neighbors, following a path which can be resolved into two com-

ponents, each parallel to one of the arrows sketched. With varying direction of the stress which is deforming the mass, the ratio between the components into which the movement of cube 12 can be resolved varies so that their resultant conforms with the direction of the stress. Here the important thing is that whereas each cube of row 3 moves in a direction parallel to its vertical axis but does not move in a direction parallel to either of its horizontal axes, cube 12 moves simultaneously in directions parallel to its two inclined axes which are parallel to the plane of the paper.

461. Slip May be in Three Directions.—Carrying the same thought a step further, if the pressure in addition to being thus oblique to two of the axes of cube 12 were in addition oblique to the plane of the paper, and therefore oblique to all three of these axes, this cube could now move obliquely to the paper, slipping simultaneously on all six of its faces, and hence in a direction parallel to none of its axes, a direction, therefore, the resultant of three movements each parallel to one of those axes, and in any ratio whatsoever, so that whatever the direction of the stress, the cube has the power of following it.

It is only for ease of explanation that I have spoken of these sliding crystalline masses as cubes. If the slip planes are octahedral, as in austenite, gold, silver, and copper, or of the {211} trapezohedron as I infer in §445 that they are in ferrite, the cubes of Fig. 65 should be replaced in the imagination by octahedra or trapezohedra.

Moreover, because the octahedral planes follow four directions and those of the {211} trapezohedron twelve, there may be slip along any number of sets of planes up to four or twelve.

Hence, in fine, a mass of any shape can in our imagination be squeezed through slip into any other shape, provided only that the stress is fit in direction and strong enough, and that the power of slip is not meanwhile exhausted.

462. Evidence that slip bands represent crystalline slip, as distinguished on one hand from twinning, and on the other hand from fluid motion. The optical methods, which in the case of transparent substances testify to twinning or to the persistence of the initial orientation through plastic deformation, are not readily applicable to metals, and hence the evidence is almost necessarily circumstantial. The most important of it is of two parts: first, the properties of the slip bands themselves, their straightness and their contours as observed directly in cross-microsections and as inferred from their optical behavior under oblique illumination and under rotation; and second, certain other indicative phenomena which form natural consequences of one or another supposed mechanism of deformation, such as the regularity of the etching and compression figures created after deformation, and the influence of that deformation on the chemical and physical properties. Of these lines of evidence only this last is of weight for distinguishing between slip and twinning.

Let us here take up these properties of the slip bands, deferring the

other indicative phenomena to the general discussion of the nature of plastic deformation in Chapter 27.

463. The Straightness and Geometrical Directions of Slip Bands.—We have seen (§441) that under favorable conditions, for instance in copper and lead which have been but slightly deformed, the slip bands instead of following the direction of the chief stress, may follow the crystallographic planes very accurately; that they change direction sharply at the grain boundaries; and that in both austenite and ferrite, as well as in other metals, they may follow the intersections of definite crystallographic planes (§§443 to 449). This certainly tends to show that they represent motion along these planes, but this motion might be either slip or twinning. It is true that their straightness and their geometrical relations may be replaced by curvature and irregularity. This, as we shall see in §498, may on one hand represent a different kind of mechanism, fluidity for instance, or on the other hand it may result from the irregular stepping back and forth from one to another of two conjugate sets of slip planes, or the deflections of initially straight slip bands by later irregularly spaced slip along conjugate planes.

464. The Profiles of Slip Bands. What They should Look Like.—(See §418.) A slip band as seen under vertical light is simply a black band in a bright surface. Its blackness implies only that the surface here is not horizontal, and hence does not reflect the light vertically back into the microscope. Such an inclined band might be a side of a rounded fold representing fluid motion, such as *k'cf'c'dfk* of Fig. 72; or it might be a crystallographic plane riser such as *h'k* or *nk*, in which case it might represent either true crystalline slip or a twin lamella; or it might be an assemblage of indistinguishably small steps, darkened by the sum of their risers. Let us see to which of these the actual appearance of a microsection cut across the slip bands best corresponds. If slip occurred as regularly as is shown in Fig. 64 by means of equidistant steps of equal height, then when seen in cross-section all treads should be level, retaining the direction which the surface had before deformation, and all the risers of any one set should be straight and parallel, intersecting the treads sharply and at a fixed angle.

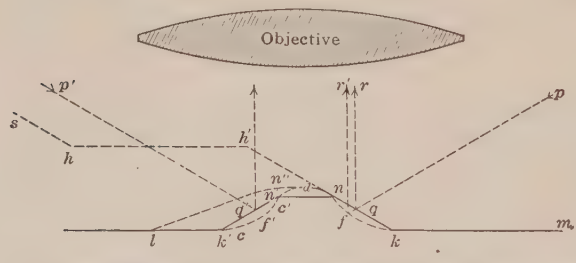


FIG. 72.—Ridges and furrows under oblique illumination are bright in two positions 180° apart.

But the mechanism of Fig. 64 does not imply such equidistance and equal height. If, for instance, between the well-marked risers *c'g* and *d'm* there were many very minute steps irregularly spaced, the line *gd'* instead of being level would look curved, though in fact it would consist of a succession of unequally spaced steplets so small as to escape detec-

tion. By like process the riser $d'm$ would be curved if it represented not a single step but an accumulation of many undetectably small steps irregularly spaced and of variable height. Under these conditions the profile of such a surface would lack all the characteristics which I have given above, for the apparent treads, being an accumulation of irregularly spaced sharp crystalline treads and risers, would not look either straight, parallel to the original direction of the surface, or parallel to each other, and the same would be true of the risers, which, moreover, would not intersect the treads sharply or at a constant angle.

Moreover, when seen in plan the surface, even under these conditions, might well have the usual look, dark slip bands alternating with bright surfaces, for under vertical light the risers are black simply because they do not reflect the light back into the tube of the microscope, whereas a tread looks bright because it does. Under our supposed conditions, that which looks in profile like a curved riser would still be dark relatively to the treads, because it has a larger proportion of black riser surface to bright tread surface than there is in what looks like a curved tread between it and the next apparent riser.

465. Secondary Steplets.—The little steps which we have been considering may be called "primary" because they result from faulting the original polished surface by slip along planes such as cc'', dd'', etc. of Fig. 63. They have horizontal treads. After a slipping which has created them, secondary slip may occur along conjugate planes such as ff', ll', and mm' of Fig. 73, breaking each of these steps up into secondary steplets, the treads of which are not horizontal. Under vertical light the whole aggregate of such steplets from c' to g' would look black, because both the riserlets and the treadlets are inclined and hence cannot reflect the light vertically.

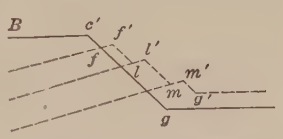


FIG. 73.—Secondary slip breaks the primary riser into secondary treadlets and riserlets.

These treadlets, instead of running north of east as in Fig. 73, might run south of east, differing from the riserlets only in being less steep.

The widening out of the slip bands which occurs under reversals of stress into broad bands may be caused in some such way.

The breaking up of the relatively high risers formed early in the deformation, such as are shown in Fig. E of Plate 25, into smaller ones as the deformation proceeds, should occur in this way, giving rise to inclined treadlets, except in the rare case in which the conjugate slip planes happen to be parallel to the initial polished surface.

466. What They do Look Like.—Turning, now, to the actual profiles, the regularity of even the most regular, those of Fig. E of Plate 25, is certainly far from complete. It is true that most of the treads are roughly parallel, yet only roughly, 2 for instance being much flatter than 1 and 3, and those from 5 to 8 being curved. The risers, again, are not parallel, those between treads 4 and 5, 8 and 9, and 10 and 11 being flatter than those

between 5 and 6, and between 9 and 10. Many of the risers intersect the treads sharply, yet some do not, for instance that on the right of 8.

The order of regularity here is somewhat less than that of the profiles of the Neumann bands shown in Figs. A to D of Plate 36. Whether the slip bands fall short of the Neumann bands in regularity more than can readily be explained by their smaller scale, may be a matter of opinion. It is hardly fair to point out that the Neumann band profiles look more regular on small than on large magnification, in Fig. A than in Figs. B, C, and D, for our present question is as to the effect not of magnification but of absolute size on the apparent regularity. A certain amount of rounding off of profiles by such processes as surface flow, splintering, and abrasion is to be expected, and this should have a greater effect on the apparent regularity of the profiles of the small slip bands than on that of the much greater Neumann bands. Indeed, that it has been possible to avoid obliterating altogether the extremely minute features of these profiles is the thing to wonder at.

It is because it is so hard, in the very nature of the case, to preserve the details of so minute a profile that these actual micrographs reinforce but slightly the other indications that the deformation is crystalline in its nature. But, as we shall see in §616, the fact that these risers reach so appreciable a height in the early stages of deformation tends to show that the slip bands represent slip and not twinning.

Were such profiles and such crystallographic directions the rule, they in themselves would incline us strongly to hold that strictly crystalline motion was the rule. Unfortunately the usual degree of regularity of profile is very much less. Thus Rosenhain's other "sectional view" is about like Figs. H and I of Plate 25, and the general appearance of those I have seen under the microscope is so much more like Figs. F, H and I, of Plate 25, as to suggest that Fig. E of Plate 25 may possibly represent Neumann bands and not slip bands. In my profiles on Plate 25 the nearest approach to the steps of Fig. 64 is at 1 and 2 in Fig. H and at 1 in Fig. I. But instead of horizontal treads alternating with parallel risers we here have sawteeth, with both directions strongly inclined to what was apparently the original direction of the surface. Though at 1 and 2 in Fig. H we can imagine a rough parallelism between the treads and also between the risers, under great deformation most of the individual salients and depressions are such as would be expected to result from fluid motion, being either wholly irregular or rounded and warty like the knob 1 in Fig. G, or the hollow 1 and the serration 2 in Fig. F. The actual deformation in these two cases was much greater than that which the micrographs themselves suggest.

In short, the sharpest and most regular of the slip band profiles as seen in cross-section in Fig. E of Plate 25 are consistent with the theory that the motion in these cases is preponderatingly and perhaps purely crystalline though they hardly add to the other evidence to this effect; but the far more usual rounded and irregular ones, though they are capable of being explained thus, yet in themselves suggest a very large degree of fluid motion.

467. General Considerations as to the Visibility of the Slip Bands under Varying Conditions of Illumination.—In the simple arrangement of Fig. 64, with its horizontal and hence bright treads, the apparent width of the risers, which are black because inclined, varies as the sine of their angle of inclination, being nil when they are vertical or undercut as in Figs. 74 and 75. If, for this simple arrangement, we substitute that imagined in the last section, an irregular spacing of the slip bands so as to suggest the curving of the treads and to give apparent variations in the directions of the risers, it must still result that under vertical light the primary treads, whether large or small, are bright and the risers dark, because by assumption these apparent curvatures are only assemblages of short treads and short risers, the deviation from the scheme of Fig. 64 being only in the regularity of the size of these elements and not in their direction. This is a point of the first importance. The secondary slip would give rise to inclined treadlets which would be dark under vertical light, but the fractions of the original primary treads would remain bright.

FIG. 74.—Slip along vertical planes.

FIG. 75.—Slip along undercut planes.

In any one grain there are 12 possible directions for the slip planes if they follow the 211 planes truly. Yet because the crystalline axes of the various grains form all sorts of angles with the surface of the specimen, the risers and the secondary treadlets may have any inclination to the original surface and hence to the primary treads, from being almost parallel to being at 45° as in Fig. 64, or at 90° as in Fig. 74, or undercut at any angle as in Fig. 75. An outward slope would naturally be expected after tensile deformation, an undercut after compressive. If, as sketched in Figs. 73 to 75, the primary risers follow one common set of crystallographic planes, they will be parallel in any one grain and so, under like conditions, will the secondary treadlets. At least the risers which look parallel on the surface of the specimen are also in general parallel in space, that is, they have the same inclination to the vertical.

468. Visibility of the Members of the Steps under Unvertical Light. The Light Plane.—Under horizontal illumination, as in Fig. 64, all primary treads must be black, and all risers must be bright which simultaneously are inclined at 45°, so that the angle of incidence[1] LNg equals that of reflection vertically into the microscope, QNc', and also are normal to the vertical plane in which the light and the axis of the microscope lie. For brevity this may be called the "light plane."

Under any light other than vertical the primary treads must always be black, but the other members may be black or bright according to the conditions. Moreover, whereas under vertical light the presence of slip bands is recognized by the contrast between their black risers and the bright

[1] For the present purpose it is convenient to designate as the "angles of incidence and of reflection" those between the light ray and the surface, instead of those between that ray and a line normal to the surface.

primary treads, under any other light the greater part of the surface is black, because those primary treads make up most of the surface, and the presence of slip bands is indicated by bright bands, which may be either risers or secondary treads. The conditions which permit either of these to become bright by reflecting the light into the microscope are that its plane is both normal to the light plane and inclined so that its intersection with that plane makes the same angle with the light and with the vertical, so that the necessary equality between the angles of incidence and reflection shall result in reflecting the light vertically. For instance in Fig. 72 the plane of the riser *nk* must first be normal to the light plane in which *pq* and *qr* lie, and second its intersection *nk* with that plane must make equal angles *pqk* and *rqn* with the light *pq* and the vertical *qr*.

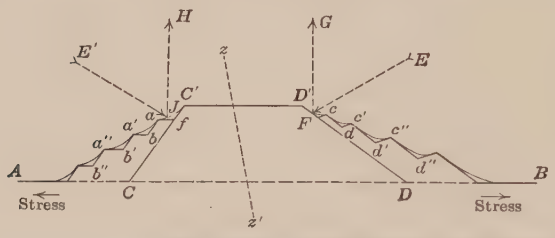


FIG. 76.—Crystallographic plane risers of slip bands, when smeared with amorphous metal, may behave optically like rounded foldings.

A curved fold, on the other hand, such as *k'cf'c'dnfk*, provided that its generatrix is normal to the light plane, could reflect vertically the light from a wide range of inclination, the part *f*, for instance, reflecting light from *pq*, and the crest *d* reflecting vertical light.

If, instead of a curved fold such as that of Fig. 72, there is a succession of curved risers such as those of Fig. 76, then any one of these, such as *aC'*, which is to be bright and hence visible, must first have a generatrix normal to the light plane; second, must have a part about *J* which is so flat as to be the equivalent of a plane, in view of the lack of strict parallelism in the light rays; and third, for a distance about *J* wide enough to cause a visibly wide beam, must make with one or another of the light rays the same angle as with the vertical, so that $E'Ja = C'JH$. For such a curved riser this principle results in the law that the curve must be cylindrical, with a generatrix normal to the light plane, and must have a section the intersection of the various parts of which with that plane make equal angles with the varying directions of the light and with the vertical.

The plane or curved surface can be made normal to the light plane by rotating the horizontal stage with the specimen about the vertical axis of the microscope.

Here, then, we have two means of distinguishing true crystalline slip, exposing crystallographic plane risers, from fluid folds, that under given non-vertical illumination few of the former but most of the latter should become bright on rotating the stage to the proper angle, and that on changing the vertical angle of the light the former should vanish but the latter should remain bright. It is true that the rays of oblique light are rarely strictly parallel, and hence the proportion of plane risers visible under given oblique illumination should be somewhat greater than has just been implied. Yet after making such allowances these general rules hold good.

Most of the slip bands in Plates 21 to 25 suggest that they represent relatively steep slip planes, because the dark bands, which are the risers, are so narrow; but the width of the black bands is so large a fraction of that of the bright ones or treads in Figs. G of Plate 22 and B of Plate 31 as to suggest that they represent very flat slip planes or secondary slip.

469. Reappearance of Bright Risers on Rotating 180° under Oblique Illumination.—When an obliquely illuminated specimen, Fig. 72, is rotated in a horizontal plane about the vertical axis of the microscope till a given set of lines becomes bright, if on next rotating it 180° these lines reappear, the natural inference is that they represent the sides of a set of liquid folds such as *cc'dnfk* of Fig. 72, rather than a set of crystallographic plane risers such as *sh* and *h'k*, for obvious reasons.[1]

The original brightness means the vertical reflection of the beam *pqf* into the tube of the microscope. This could be caused by either the plane surface *h'k* of a riser, or the rounded surface of a fold as at *f*. The reappearance of the bright lines after rotating 180°, or what is the same thing, after reversing the direction of the light from *pq* to *p'q'*, means that there is a vertically reflecting surface symmetrical with the first, that is, a plane or cylindrical surface, inclined to the first as left and right about a vertical plane, with a horizontal generatrix parallel to that of the first plane. The horizontal generatrices of the two surfaces must be parallel in order that both shall be normal to their common light plane, and they must be as right to left about a vertical plane parallel to those generatrices in order that they shall reflect vertically light which is similarly inclined.

The existence of such a surface would be frequent in the case of liquid folds, the parts *f* and *f'* of the lower curve *cf'c'dnfk* bearing exactly this relation to each other. Yet if, as might often happen, the left-hand side of the fold were throughout too flat to reflect *p'q'* vertically, as the upper curve at *ln''* is, the bright line would not reappear on rotating 180°. The proportion of cases in which it would fail to reappear should increase with the flatness of the light. Here is a test which might help greatly to decide whether these slip bands represent crystallographic planes or rounded folds.

But a crystallographic plane *k'n'* symmetrical in this way with *kn* could exist only under rare contingencies such as that the plane passing through two of the crystalline axes of the grain in question has this same symmetry, and hence is normal to the surface of the specimen.

For instance, if the plane *nk* of Fig. 72 is replaced by a cubic one inclined at 45°, then, assuming that the paper itself is vertical, one of the two other directions of cubic planes can supply a plane which matches *nk* as *n'k'* does, provided that one of the crystallographic axes is horizontal and normal to the plane of the paper, and that the two other axes are parallel

[1] Under oblique light the Beck or other reflector, or some other bright object, may mislead us by causing a false beam like *p'q'* which, on 180° rotation, will bear to the specimen the relation which the true beam *pq* bears before rotation, and will thus brighten, though more feebly, the surfaces which shone before.

to the plane of the paper and make angles of 45° with the vertical. Or, if *nk* is replaced by an octahedral plane, then any one of the three other octahedral directions can supply a plane to match it, and this matching will occur if one axis of the octahedron is vertical and the two others are horizontal and make angles of 45° with the plane of the paper.

But if the vertical axis of this octahedron is tipped to the left as at *zz'* in Fig. 76, the opposite planes corresponding to *nk* and *n'k'* will not be symmetrically inclined to the vertical, and hence cannot reflect the symmetrical light beams *pq* and *p'q'* vertically into the microscope, and hence cannot replace each other as bright risers on 180° rotation. If its upper end is tipped toward the reader, the two planes *nk* and *n'k'* can no longer be normal simultaneously to a common light plane, and hence the angle of rotation which would cause them to replace each other as bright risers is not 180°.[1] These same considerations apply to the slightly more complex actual case, in which the risers follow the 211 trapezohedral planes instead of being cubic or octahedral.

In short, liquid folds should often but not always have bright lines which reappear on rotating 180°, but crystallographic plane risers should only rarely. Here, then, we have a third criterion for distinguishing crystallographic plane risers from fluid folds, that the former should rarely reappear on 180° rotation, whereas the latter should often but not always.

470. The optical behavior of slip bands is suggestive of fluid folds rather than of crystallographic plane risers, according to the three criterions, A, B, and C, explained in the last two sections.

A. On rotating the specimen under oblique illumination the total number of bright lines which shine out in one position or another, compared with the true grand total seen under vertical illumination, gives the impression of being greater than it should be if they were crystallographic planes (§468).

B. A given set of bright lines remains visible even when the vertical angle of the light varies widely, and even when a narrow slit parallel to them lies between them and the objective,[2] which suggests that they represent curved surfaces and not crystallographic planes (§468). They indeed reach their brightest under light from one fixed angle, but this might well be true of curved folds, for instance if, in Fig. 72, the radius of curvature were longer at *f* than elsewhere.

C. In a given grain under oblique light, the bright lines which reappear on 180° rotation,[3] though indeed with a different degree of bright-

[1] This is true no matter how slight the tipping, but it is seen mentally more easily if a marked tipping is imagined, say of 45°.

[2] This is asserted by Osmond and admitted by Rosenhain (*Journ. Iron and Steel Inst.*, 1906, II, vol. 70, pp. 193–4).

[3] Rosenhain has verified this 180° reappearance, or more strictly the simultaneous shining of symmetrical surfaces, like *nk* and *n'k'*, of Fig. 72, with characteristic ingenuity by illuminating the field simultaneously with beams of two colors from two opposite directions. "Where bands of two colors appear upon the same crystal at all, it is usual to find a considerable preponderance of

ness, form a fraction of the whole so great as to suggest that they represent the two sides of a fold, of which one side is often too flat to reflect the oblique light vertically, especially because other evidence points to the existence of such folds, rather than that they represent symmetrical sets of crystallographic plane risers, especially because such symmetry could exist only in a minute proportion of the total number of grains (§469). Of course the number of such planes facing to the right should differ somewhat and might differ greatly from that facing to the left.

The importance of this matter is so fundamental that it calls for an examination more thorough than any thus far made.

471. How Suggestions of Fluidity may Arise.—Let us ask whether important suggestions of fluidity may be explained by the conception that, though slip does occur along true crystallographic planes, giving rise to crystallographic plane risers, yet these in effect act like rounded folds, first through breaking up under progressive deformation into irregularly spaced steplets, and second through being smeared over by the amorphous metal generated first by the previous polishing and hence adhering primarily to the treads, and second by the slip itself, and hence buttering the risers, being drawn down the face of each riser $C'f$, ab, etc., of Fig. 76 by the friction of the block at its left. In this way might arise the rounding of the steps here shown, which is not unlike that of the rounded steps of Fig. E of Plate 25.

Let AB represent the initial polished surface. Assume that plastic deformation has developed slip along the crystallographically equivalent planes CC' and DD', which are symmetrical about a plane running through the inclined crystallographic axis zz' and normal to the plane of the paper. Under these conditions both risers are normal to the light plane. Of these risers only DD' could possibly reflect light into the microscope, because CC' is steeper than 45°. Moreover, DD' could reflect vertically only light with the inclination EF. Thus CC' and DD' comply with the three rules already laid down for crystallographic plane risers, for only half of this pair of risers can be made to look bright in the microscope under any conditions, the brightness of that half needs light from a definite direction both horizontal and vertical, and it does not return on 180° rotation.

If, now, as is implied by Rosenhain's observation that the height of the risers decreases as deformation proceeds, each of these risers breaks up by secondary slip into many steplets, ab, $a'b'$, etc., and if we assume that the amorphous metal generated along the slipping surfaces rounds itself through surface tension as sketched, then these steps, instead of behaving as crystallographic plane risers should theoretically, would in fact behave like rounded folds.

Such a rounding of the amorphous metal seems needed to explain the bright bands which so often run between the treads and risers of slip bands

one color with comparatively few of the other interspersed here and there" (*op. cit.*, p. 195). My own observations with simple 180° rotation indicate that the proportion which thus reappears is greater than this language might suggest.

(§412). That these bands are so often brighter than the initial surface, as represented by the treads or spaces between slip bands, indicates, first, a slight tilting of the specimen so that the treads are not normal to the light,[1] and second a bright band between tread and riser which owes its brightness to being more nearly normal than the treads. Were the intersections of treads and risers perfectly sharp, no bright band could exist there. The width of this bright band is often so great as to suggest strongly that it represents the curved surface which I have imagined, the amorphous metal drawn over the nose of the step by the slip, like the surface of water over a weir.

Indeed a simple fold of the amorphous surface layer developed by the polishing may in and by itself explain the optical suggestions of fluidity. In addition a development of amorphous metal on the risers by the slip itself seems needed to explain the bright band which so often runs alongside the dark riser of the slip band when the initial horizontal surface, that of the treads, lacks the amorphous polishing layer. This is true of the slip bands not only on the unpolished fracture, Fig. B of Plate 35, but also of the etched surfaces of Plate 23. Here an etching deep enough to mark out the grain boundaries so well should have removed the whole of the amorphous polishing layer. In all these cases this bright band runs alongside some of the dark risers, though it is not so striking here as in polished and unetched surfaces, such as E of Plate 22.

472. Detailed Reasons for These Suggestions.—First, because some part of these rounded surfaces is likely to make with the vertical the same angle that it makes with the light from whatever direction, it follows that most of the slip bands will become visible by thus reflecting light vertically, on rotating the stage to the position in which the axis of the slip band is normal to the light plane. This would explain an abnormally large proportion borne by the total number of slip bands visible at the various stages of rotation, taken collectively, to the total number visible under vertical illumination. It would explain also an abnormally wide vertical range of directions of light under which any one set of slip bands should be visible. This conception requires that the profile of the smearing shall be greatly rounded.

Second, for given direction of light the width of the band so inclined as to reflect that light vertically will not be the same on the left as on the right hand, but probably narrower say on the left than on the right as here sketched. Hence one of these two sets of risers will look brighter than the other.

Third, because the planes passing through the risers on both the right- and left-hand sides of this uplift are normal to the light plane, the angle of

[1] It is this tilting that, by compensating in part for the deformation of the surface, enables the whole of the field to be in focus simultaneously. If, for instance, the treads of the slip bands in the large grain in Fig. D of Plate 22 were strictly normal to the light, then, short as the risers are, their height acting cumulatively would make it impossible to focus the whole field simultaneously. Instead the specimen naturally becomes tilted in mounting so that it is not the treads but a plane passing approximately through the noses of the steps that is horizontal.

rotation, from the position in which the right-hand side shines brightly to that in which the left side does, must be 180°, for this will bring the left side to that normality.

Fourth, a given set of steplets will become visible only in case the part of its curved surfaces inclined so as to reflect a given light vertically is wide enough to reflect a band of light wide enough to be seen. But the width of such a band on each steplet, and its collective width on the whole set of steplets, will vary with the contour of the steplets themselves, so that the right-hand set may be visible though the left is not. Hence if the specimen is lighted by symmetrical beams of light EF and $E'F'$ of different colors, the total number of slip bands visibly lighted by one will not be the same as that lighted by the other. In this way it will happen that in a given grain some of the right-hand sets of steplets are visible but none of the left-hand sets are.

Fifth, if the landings $C'D'$ are extremely narrow, as may well happen, then a succession of slip bands, each consisting of an aggregate of steplets, will look in profile like sawteeth. This agrees with my observation that only under favorable circumstances can a profile like that of Fig. E of Plate 25 be had, and that instead a sawtooth detail is far more common.

Thus in all these respects the conception of the smearing of the steps by amorphous metal explains the observed facts.

The difficulty is to admit that this amorphous smearing could be large enough and rounded enough to cause visibly wide bands of vertical light. The mere breaking up of the steps into crystallographic plane treadlets and riserlets which retain the directions of the initial ones would not have these effects, because each of these treadlets and riserlets would act optically exactly like the original. At least the resultant fractions of steps would consist of a small number of sets each of fixed inclination, and thus each set would act on the light as one whole. A true as distinguished from an apparent rounding of the surface is needed.

473. Slip Bands Imply the Localizing of Deformation.—The existence of slip bands at appreciable distance apart implies that the deformation, whatever its nature, is concentrated along certain internal surfaces, instead of being spread out evenly as in an expanding gas. There is indeed a suggestion here that the deformation is along certain definite crystallographic planes, and hence either by slip or twinning, but it is only a suggestion, for such local concentration is manifested by the rufflings in varnish, the creases in cloth, the lines of Lüders, if not indeed by the waves of any fluid. The concentration in these cases may or may not be complete, but the important thing is that the internal surfaces about which it occurs are not crystallographic planes.

474. Evidence of Complete Localization.—Instead of this supposably incomplete concentration, a complete localization of the deformation in lead has been proved by Rosenhain.[1] He finds that if, before deforming a

[1] *Journ. Iron and Steel Inst.*, 1906, No. II, vol. 70, pp. 191–192.

specimen, fine figures are set up on its surface, their parts which lie between the later formed slip bands are not changed in either size or shape by the deformation of the specimen. Such figures may be formed either by fixing by means of smoke the position of drops of moisture deposited from the breath on the surface of the specimen, or by smoking that surface and then drawing fine lines on it with a brush.

The fixity of these figures he proved by micrometric measurement of the distance between the brush lines, and by superposing stereoscopically photographs of the surface between slip bands, taken before and after deformation. Ferrite, he believes, cannot well be studied thus, because its slip bands lie so close together that there is no room for recognizable figures between them. Yet I find that in coarse crystalline ferrite, such as the A.R.M. steel and thermit steel, slip bands very far apart can be developed by gentle deformation.

That the deformation is localized in austenite is shown by the identity of Rosenhain's figuring caused by deformation with that caused later by repolishing and etching the deformed surface, showing that the etching is accelerated in the bands along which the deformation has occurred.[1]

475. The Hardening and Like Effects of Plastic Deformation Point to Slip.—Every form of plastic deformation increases the strength, hardness, brittleness, volume, and solution pressure. These effects are a more natural consequence of the generation of amorphous metal, such as would naturally result from crystalline motion, than of fluid motion (§§513–518). But though this gives us an additional indication that there is crystalline motion, it is easily compatible with the belief that there is also much fluid motion.

476. The Failure of the Slip Bands to Reappear on Repolishing and Reetching.—If the slip bands should prove, on further study, to differ radically from the Neumann bands in this respect, so that their failure to reappear thus is due to something more than the undetectable narrowness of their etching traces, that would tend to show that they do not represent twinning, and hence by inference that they represent slip (see §617).

477. Summary of the Evidence as to the Nature of Slip Bands.—On one hand, the crystallographic direction of certain slip bands, and to a less degree their straightness and the occasional sharpness of the intersection of their treads and risers as shown in Fig. E of Plate 25, and the complete localization of the deformation along certain planes, tend to show that the slip bands represent slip along crystallographic planes, with the development of fairly sharp treads and risers. On the other hand, the curvature and absence of crystallographic orientation of many of them even after slight deformation, their extreme crookedness and irregularity after deformation, and the three optical tests rather suggest that very many of them represent a different mechanism of deformation, such as the formation of fluid folds.[2]

[1] *Journ. Iron and Steel Inst.*, 1906, No. II, vol. 70, pp. 212, 213, Figs. 12 and 13.

[2] Osmond, whose opinion deserves the very greatest weight, took a view much like this (Osmond and Cartaud, "The Crystallography of Iron," *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71,

Yet even this evidence is capable of being reconciled with the hypothesis that they all represent strictly crystalline motion.

The great hardening and other effects of plastic deformation point to crystalline rather than to fluid motion, yet they do not imply that the motion is solely or even chiefly crystalline. They connect themselves rather with slip than with twinning, and they are the only evidence thus far studied, unless it be the reported failure of slip bands to reappear on repolishing and re-etching, that tends to show by which of these forms of crystalline motion the slip bands are caused.

The foregoing, be it understood, is only an incomplete summary of the evidence. We shall see in Chapter 24 that twinning probably does not contribute to plastic deformation, and that hence, intergranular deformation being slight except in the early stages, slip and fluid motion remain by elimination as the chief mechanisms of this deformation. We shall then see in Chapter 27, that the regularity of the etching pits and of the etching tint even after great deformation indicates that this deformation must have been primarily crystalline, so that fluid motion must be of minor importance, and slip the chief mechanism of plastic deformation.

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pp. 447 to 449 and 461). He interpreted only the straight slip bands as representing crystalline motion (pp. 447 and 461), and held that the slip planes, which he called planes of translation, were only those of the easiest translation. When the power of translation along those planes has been drawn on, the translation may thereby become relatively easier along a second set. As deformation thus proceeds, the mass becomes in part decrystallized, and folds replace translation along the crystallographic planes. In particular he says of his own silhouettes in ferrite, that, though their straight elements prove the existence of certain crystallographic planes, "the curved lines dominate considerably; therefore the translation is difficult, and the greater part of the deformation seems due to another mechanism." He seemed to reject the hypothesis that the curvature may be due to either the curved integration of many short straight tangents or to many minute faultings of the slip band after its formation. In 1904, writing with Fremont and Cartaud, he had referred the slip bands wholly to fluid movement (*Rev. de Metallurgie*, "Les modes de deformation," *Memoires*, 1904, vol. 1, p. 11).

CHAPTER 19

DISCUSSION OF SLIP BANDS (*Concluded*)

478. Introduction.—In §§479 to 488 I consider the effects which may be expected to flow from the lack of registry between the slip planes on the opposite sides of grain boundaries. In §§489 to 497 I offer and discuss evidence bearing on the existence of these expected effects, and in §§498 to 504 I consider the curvature of slip bands.

479. Effect of Grain Boundary Disregistry on the Propagation of Slip.—The change of orientation at the grain boundaries, implying that the slip planes on the two sides of the boundary do not register, or in short implying disregistry, might be expected to have two distinct effects on the propagation of slip: first, a direct resistance to it, and second, an indirect and cumulative effect through the generation of amorphous metal there, as I will now show.

480. Direct Obstruction.—Ignoring for the moment the movement in grain *BDPQ* of Fig. 65, p. 337, of the rows other than 3, we see that the movement of this row within this grain is opposed only by the friction or cohesion along its four vertical sides parallel to the motion. But the movements which must occur within grain *MKLN* in order to permit cube 3*C* to thrust itself down like a tongue, yet without loss of continuity, might be supposed to meet much greater resistance, first, because only one component of the vertical thrust is available to cause this movement, the other component pressing cube 13 down against 24 and other cubes in like manner down upon those below them, and thus increasing the frictional resistance; and second, because the extent of the surfaces along which slip must needs occur is very much greater in the lower than in the upper grain. For instance, if we suppose that the gusset block 7 is broken by the descent of cube 3*C* as sketched on a larger scale in Fig. 66, and that its fragment *nopq* slides down to the position $n'o'p'q'$, there is, in addition to its sliding along the planes op' and nq', the transverse sliding of its face across the end of cube 3*c* to the left, to the position $o'n'$. But even this cannot satisfy the conditions, because it would leave an empty gusset nen', and the fact that continuity is not broken and the density is not changed permanently implies that no such empty space forms. This implies even more complex movements, and hence even more extensive surfaces, the friction along which opposes the descent of cube 3.

Moreover, the breaking of gusset block 7 has to overcome the cohesion of its parts, which we may well believe to be much greater than that between adjacent cubes along slip planes, which are almost by definition planes of low cohesion.

Thus the fact that the slip planes in the grain on one side of a boundary do not register with those on the other side, or in short the *intergranular*

disregistry, obstructs the passage of slip across the boundary, and may be likened to breaking joint in masonry.[1]

481. Because of Disregistry, Slip should Generate Amorphous Metal at the Grain Boundaries.—The breaking of the gusset block 7 of Fig. 65 and the complex shearing movements in the grain *MKLN* may well break up an appreciable quantity of the crystalline metal somewhat as polishing does, and, destroying its crystalline nature, bring it to the amorphous state, which, as explained in §508, is probably much stronger than the crystalline state. If absolutely no empty spaces arise, then no matter how minute we conceive the crystal units to be which accommodate themselves to motion across such a disregistering boundary, some of them must be broken up. You cannot pile bricks disregisteringly without either leaving some crevices of the order of size of a brick or else filling those crevices with broken bricks. So with the contact of disregistering systems of crystal units: either crevices of their order of magnitude, or crushing, which implies amorphization. This generation of amorphous metal in this disturbed region would in itself add to disregistry in opposing the propagation of slip past the grain boundaries. After such amorphous metal has once formed along the grain boundaries, any further slip has to overcome not only the disregistry but also first the resistance which this amorphous metal opposes to slip because of its lack of any slip planes, any planes of low cohesion, and second any excess of strength of the grain boundary region to whatever cause it may be due. Here the throw is so great that the disregistry cannot be accommodated by the elastic movement of the constituent particles.

482. End-support for the Slip Planes.—A tensile test piece is composed of innumerable grains with all manner of orientations, and therefore with all manner of inclinations to the existing stress, some more and some less favorable for resisting that stress. Each grain has various sets of slip planes, varying in unit shearing strength. Shear may be expected to occur the more readily along a given plane if that plane registers exactly with those of the adjoining grains, so that the shear extends readily across the grain boundary. In this sense disregistry acts as an *end-support* to reinforce the shearing resistance along any disregistering plane to the existing stress. So, too, not only will the slip planes of one grain differ from those of another in the degree of support which they receive from this disregistry, but even among the slip planes of a given set and in a given grain some will thus receive more support than others, the support thus given the longer planes being less per unit of their area than that simultaneously given the shorter ones. In short, even among the slip planes of a given set in a given grain, not only will some have less unit shearing strength than others but also some will have less end-support than others.

483. Plural Causes of Variation in the Resistance to Slip.—Thus in

[1] Rosenhain ("Physical Metallurgy," 1914, p. 275), though he recognizes this impediment which grain boundaries offer to the propagation of slip, does not consider the far-reaching effects which it suggests.

addition to the variations in the direction and intensity of the external stress in the various parts of the test piece, there are inconceivably numerous variations in the inclination of the various slip planes toward that stress, variations in the unit shearing strength of the various sets of slip planes as sets, and of the individual planes of each set, and in the end-support which the individual slip planes receive. Of all the slip planes in the test piece there will be one the least favorably disposed, in that it has the highest ratio of the component of the external stress which tends to cause shear along it to its net unit resistance resulting from its intrinsic unit strength and its end-support. And even in that least favorably dispose plane there is one crystal unit which is less favorably disposed than any of the others, the least favorably disposed crystal unit of the whole test piece.

484. The End-support Hypothesis Applies also to Twinning.—We shall see in §619 that the end-support which each plane is thus held to receive where it abuts against the boundary of the adjoining grain should be expected quite as much if the movement is assumed to occur not by strictly parallel slip without rotation, but by rotary movement along certain definite crystalline planes, that is, by twinning. But for simplicity we may discuss the matter with slip alone in our minds.

485. The First Slip Band and the Elastic Limit.—As the external stress increases progressively, the intensity of its component along this least favorably disposed unit (§483) will at last surpass the effective resistance of that unit, and slip will occur. In that this unit slips and thus fails to stand up to its load, it throws part of that load on its neighbors already almost overloaded with their own share of the stress.

The yielding of this first crystal unit is the passing of the elastic limit, for the elastic limit by necessary implication is the least unit stress which impairs perfect elasticity, in the sense of the exact return to the initial dimensions. This displaced unit will resist the exact return of the test piece to its initial dimensions, and will thereby cause an immeasurably small permanent stretch. Further increment of the external stress will lead to like overloading of the next least favorably disposed unit, and thus a process starts which ends only with rupture.

The fact that this stretch when it begins is immeasurably small, and hence that from the nature of the case the beginning of permanent stretch can never be detected, ought not to suggest that it begins with the first application of the stress however minute, and hence that there is no finitely large elastic limit (§383).

The first immeasurably small stretch begins, not with the first finite stress, but, as has just been shown, when the stress just overmasters the effective strength of the least favorably disposed unit. Though this first permanent set is immeasurably small, the stress which causes it may yet perhaps be estimated approximately by extrapolation.

The appearance of the first slip band on a polished surface means that the slip along one of the least favorably disposed slip planes among those

which crop out on that surface has caused a visibly long and wide step on that surface. This in turn implies that the elastic limit has been exceeded materially, for the true beginning of the very first slip, which marks the true elastic limit, must precede materially the growth of that slip in two directions, length and width, to such size that it causes a visibly wide slip band, and this first slip of all is very unlikely to occur in the specific minute field which happens to be under our microscopic examination. For simplicity we may pass by the improbability that the first slip of all should happen to occur on the surface and in this respect be detectable, because the surface particles may, indeed, be less effectively supported by their neighbors than those in the interior, so that the first slip may occur on the surface and not in the interior of the specimen.

486. Consequences to be Expected from the Progressive Accumulation of Stronger Amorphous Metal about the Grain Boundaries (§481).—This, like every other form of boundary strength, may be expected to give to the slip planes an end-support, the effectiveness of which should be greatest in the nearest parts of those planes and should be inversely as the square root of the area of those planes, with the result of strengthening the lateral relatively to the median planes, and thus of tending to concentrate deformation and rupture in the middle of the grains. It should further tend to cause end curvature of the slip bands, and to make fine-grained metal, with its more abundant grain boundaries, stronger than coarse-grained, an effect which in turn should help to explain the increase of strength, elastic limit, and hardness with increasing carbon content, and the strengthening influence not only of the grain-refining and sorbitizing processes, but also of the grain-refining or sluggardizing elements such as manganese and vanadium.

The progressiveness of the accumulation of this stronger boundary metal as deformation proceeds, and the resultant progressiveness of the strengthening of the grain boundaries, should lead to an increasing avoidance of those boundaries, and an increasing excess of strength of fine- over coarse-grained metal, with increasing deformation.

Because such amorphous metal should become crystalline on high heating, the tendency of rupture to avoid the boundaries should cease at high temperatures.

Of these expected consequences of the imagined boundary accumulations of amorphous metal, the end curvature of the slip bands will be considered in connection with the causes of their irregularity in §503; the tendency of the end support to concentrate deformation in the grain centers in §487; the progressive increase of this tendency as deformation proceeds in §488; the corresponding progressive increase in the excess of strength of fine- over coarse-grained metal in §§489 to 496; the help which grain size gives in explaining the changes in the tensile properties caused by increasing carbon content and by sorbitizing in §§536 to 540; and the cessation at high temperatures of the avoidance of the grain boundaries in §§689 and 711. The

explanations hitherto offered of the avoidance of the grain boundaries by rupture will be considered in Chapter 28.

487. The median slip planes should be effectively weaker than the lateral ones, so that slip and rupture should occur by preference along the median slip planes, and should avoid the lateral ones.

Imagine, for instance, that slip or rupture is about to cross grains 2 and 5 of Fig. 77. If it passes down along the border of grain 5, as at *CD*, it will have to cross twice as many grain boundaries as if it passed down through the middle, as at *AB*. Hence *CD* is avoided as the path of greater resistance. Or, more generally, the end-support which a given slip plane receives, per unit of its total area, is inversely as the square root of that area.

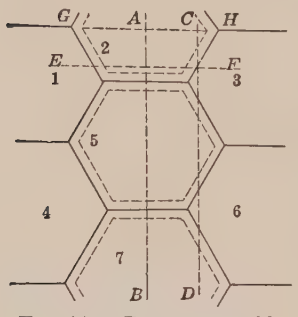

FIG. 77.—Rupture avoids grain boundaries.

But the planes with the greatest area are precisely the median ones. Moreover, in any given slip plane the end-support per unit of area is least effective at its middle and greatest at its ends. These considerations tally with Humfrey's observation that the slip bands begin at the grain centers and propagate toward their terminal boundaries.

Thus the effectiveness of the end-support should increase as we pass from the grain center toward either the lateral or the terminal boundaries, and should reach its maximum at those boundaries themselves.

488. Increase of the Avoidance of the Grain Boundaries as Deformation Progresses.—Section 407 shows us that deformation, in its early stages, is often concentrated along if not actually in the grain boundaries, giving rise to the uplift of individual grains along their contacts with their neighbors. This contrasts with the general observation that rupture, which represents the path which deformation takes when it is at its maximum, tends to avoid those boundaries, and still more with the evidence which I give in §694 tending to show that rupture not only avoids the boundaries but actually seeks the centers. This contrast between the tendency of deformation to be intergranular in its incipiency and its tendency to be centrally transcrystalline at its end naturally suggests such a progressive change from this initial to this final tendency, with progressing deformation, as is called for by my hypothesis of a progressive boundary accumulation of strong amorphous metal. This suggestion remains to be checked by direct observation. For that matter I have to admit that my evidence as to this contrast itself is not as abundant as it should be.

489. Outline of Evidence that the Excess of Strength of Fine- over Coarse-grained Metal Increases with the Deformation.—Though coarseness of grain is known to cause brittleness (§523), I have not found direct recorded evidence that it causes the weakness under quiescent stress which we should expect it to through its lessening the number of grain boundaries which

oppose deformation. Indeed, so far as I have noticed, the opinion of the best informed is that it has little effect in this respect.[1]

Some of the cases of brittleness which readily occur to us are irrelevant. Thus low-carbon steel coarsened by heating after overstrain,[1] and Ruder's coarsened silicon steel,[2] are reported to be so brittle that they break in pieces when dropped on the floor from a slight height, but in the latter case we know, and in the former we may suspect that the coarsenesss is accompanied by intergranular weakness. Hence neither case assures us that the grains proper, as distinguished from the intergranular union, are weakened by coarsening. The extreme brittleness of Osmond's crystal (§523) is evidently not intergranular but trans-crystalline.

TABLE 18.—GRAIN COARSENESS WEAKENS AND EMBRITTLES (SEE P. 362)

Material	Grain size, sq. mm.		Ratio	Tensile strength, lb. per sq. in.		Yield point, lb. per sq. in.		Yield point ratio, per cent.		Elongation in inches, per cent.			Contraction in width		Sclerometer hardness	
	Fine	Coarse		Fine	Coarse	Fine	Coarse	Fine	Coarse	Fine	Coarse	In.	Fine	Coarse	Fine	Coarse
Silicon steel 2.5 per cent. Si	....	40	.....	64700	42950	50240	39760	77.7	92.6	....	5.5	2	....	....	....	....
	....	67	.....	64910	46500	52150	46500	80.3	100.0	....	3.5	2	....	....	19	15
	....	36	.....	64540	47750	48770	43160	75.6	90.4	11.5	8.0	2	3.4	....	19	15
	....	32	.....	68120	46530	52310	42750	76.8	91.9	10.5	6.5	2	5.7	2.7	19	15
Average	.0155	40	1:2580	65570	45940	50870	43040	77.6	93.7	11.5	5.9	2	4.5	2.7	19	15
Excess (+) or deficit (−) of the properties of the coarse grained relatively to those of the fine grained in percentage of the latter.	+2480		.....	−30.0		−15.4		+20.7		−48.7			−40.0		−21.0	

Note to Table 18.—Sheets of Hadfield's silicon steel 0.013 in. (0.330 mm.) thick, some of them in their initial moderately fine-grained state and some of them coarsened by heating, and kindly given by Mr. W. E. Ruder, were cut into strips ⅝ in. wide and 6 in. long. They were then filed down for 2 in. in the middle of their length to a width of 0.58 in., and tested tensilely. The grains of the coarsened specimens were much longer than wide, as noted by Ruder (*Trans. Amer. Inst. Min. Eng.*, 1913, vol. 47, p. 575).

The rupture of the coarsened specimen passed squarely across the bodies of the grains, and not along their boundaries. This could readily be seen because of the great coarseness of the grains, which were in many cases as much as ½ in. in length. This steel contains about 2½ per cent. of silicon and 0.05± per cent. of carbon.

[1] Charpy, *Rev. de Metallurgie, Memoires*, 1910, vol. 7, p. 656. Of the coarseness of grain caused by heating low-carbon steel to about 650° after it has been overstrained he says: "This coarsening of the grain, as we know, has no considerable effect on the properties as determined by the usual tensile test, but on the other hand it affects the fragility enormously, so that the resilience as measured by the notched bar impact test is practically nil when a certain degree of coarseness is reached."

Heyn (*Journ. Iron and Steel Inst.*, 1902, No. II, vol. 62, p. 75) found a wire rod about ¾ in. in diameter so brittle that it broke in two on falling to the floor, though its composition and "the tensile tests failed to throw any light upon the cause." He found that there was little difference in tensile strength between brittle and ductile soft steel (*idem*, p. 76); that refining coarse grain might not remove brittleness; that increasing the grain size of copper more than 1,500-fold, from 66 to over 100,000 μ^2, might lessen its bending capacity by less than one-fifth (*idem*, p. 103); and on the other hand, that long heating at between 700° and 890° might cause great coarseness without other form of brittleness than that shown in the bending test (*idem*, p. 109).

[2] *Journ. Industrial and Engineering Chemistry*, 1913, vol. 5, p. 456.

TABLE 19.—SOFTENING EFFECT OF GRAIN COARSENESS

On annealing ferrite after differential deformation, the progressive decrease of the Brinell hardness caused by increasing deformation is interrupted in the critical region of grain coarsening, as shown by broad-faced type.

Material	Steel No. 2.1, C = 0.05			A.R.M. steel, No. 0, C = 0.02								Steel No. 2.1, C = 0.05						Steel from Prof. Sauveur, C = 0.08		
Method of deformation	Tension on tapered bar			Compression of single wads								Cold rolling a piece tapered in thickness								
Specimen	1			2				3				4			5			6		
Line	Stress in 1,000 lb. per sq. in.	Hardness	Grain size, μ^2	Stress in 1,000 lb. per sq. in.	Hardness: Observed	Hardness: Per cent. of initial	Grain size, μ^2	Stress in 1,000 lb. per sq. in.	Hardness: Observed	Hardness: Per cent. of initial	Grain size, μ^2	Reduction, per cent.	Hardness	Grain size, μ^2	Reduction, per cent.	Hardness	Grain size, μ^2	Reduction, per cent.	Hardness	Grain size, μ^2
1	36	85	....	38.6	88	111	7,000	60	68	103	12,500	4	90	.....	16	..	900	6	.....	.....
2	39	**83**	**3600**	41.6	89	112	7,100	62	66	102	10,900	....	90	.....	...	85	2900	(13)	**79**	**19,400**
3	42	86	....	44.6	89	112	8,500	64	67	103	12,800	....	87	.....	....	85	....	(17)	82	5,000
4	45	82	....	46.1	88	111	8,000	66	67	103	11,000	11	**85**	**40,000**	....	**85**	....	21	..	.....
5	....	85	....	47.6	87	110	10,900	68	67	103	13,300	....	90	.....	25	83	.....	....	77	.....
6	....	85	....	49.1	86	109	11,000	70	67	103	12,400	....	90	.....	....	80	.....	....	78	.....
7	....	83	....	50.6	82	104	10,700	72	67	103	15,200	15	90	.....	....	80	....	31	..	.....
8	....	82	....	52.1	83	105	13,500	74	67	103	12,500	....	88	.....	31	80	....	(33)	79	1,000
9	....	80	....	53.6	83	105	11,600	76	**68**	**104**	**12,300**	....	84	.....	....	79	.....	....	76	.....
10	....	80	....	55.1	79	100	12,700	78	64	98	24,100	20	84	5,700	....	74	....	(37)	75	3,600
11	....	80	....	56.6	76	96	12,100	80	61	94	17,300	21	84	.....	36	80	....	39	..	.....
12	**57**	76	400	59.6	79	100	13,300	82	61	94	35,500	....	81	.....	....	78	1500	....	..	.....
13	....	..	....	....	....	....	.....	....	....	....	.....	....	81	.....	....	79	.....	....	.....	.....
14	....	..	....	....	....	....	.....	....	....	....	.....	25	85	.....	40	76	....	....	.....	.....
15	....	..	....	....	....	....	.....	....	....	....	.....	....	80	.....	....	81	.....	....	.....	.....
16	....	..	....	....	....	....	.....	....	....	....	.....	26	78	2,200	....	75	.....	....	.....	.....
17	....	..	....	....	....	....	.....	....	....	....	.....	....	..	.....	....	76	.....	....	.....	.....
18	....	..	....	....	....	....	.....	....	....	....	.....	.....	.....	.....	42	71	700	....	.....	.....
Annealed at 1,050°± before deformation.	....			....	79	100	.....	....	65	100	.....	....	85	1,300	....	85	900	....	87	400
Annealed at 1,050°± after deformation.	....			....	....	....	.....	....	....	....	.....	....	78 to 82	.....	....	77 to 82	.....	....	77 to 79	.....

Note to Table 19.—Specimen 1.—A tapered tensile test piece was prepared from steel No. 2.1 of 0.05 per cent. of carbon. This was then subjected to tensile stress which gave approximately the stress indicated per square inch of the initial section of the various parts here represented.

After this straining it was heated for 2 hours to between 880° and 890° for the purpose of coarsening the parts which had undergone the critical degree of overstrain.

Specimens 2 and 3.—Twelve rectangular prisms of the A.R.M. steel No. 0 of 0.01 per cent. of carbon, about 0.3 in. × 0.35 in. × 0.5 in., and twelve 0.25 in. × 0.25 in. × 0.4 in., cut from the same bar as received from the maker, were subjected to the pressures indicated, while standing on end. The ends of the prisms were greased to lessen the tendency to bulge. In fact very little bulging occurred, except under the greatest pressures. Under these the prisms became somewhat rhomboidal.

The first twelve were then heated for 19 hours to between 680° and 710°, and the second for 17 hours to between 700° and 750°, in order to coarsen those which had undergone the critical degree of overstrain. Their hardness was then determined in the usual way, and the grain size was measured from photomicrographs. The change in the size of the prisms was necessitated by an unforeseen detail.

The Brinell hardness of the second twelve is less than would be given by specimens of normal size, and is useful only for comparing the members of this series. So, too, the hardness of the two lots is not directly comparable, and hence is reduced to percentage of the initial hardness as determined on prisms of these two sizes.

Specimens 4, 5, and 6.—Tapered plates were rolled cold between horizontal cylindrical rolls

so as to give the degree of reduction indicated. They had been preheated, Nos. 4 and 6 for 15 minutes to between 1,000° and 1,025° and No. 5 to about 1,070°. Each of them was then air cooled.

They were then rolled as indicated in the table, and then coarsened. In this coarsening Nos. 4 and 6 were heated during 2 hours from 300° to 600°, then held 40 minutes at 820° and furnace cooled. No. 5 was coarsened for 2 hours at 870° and furnace cooled.

The hardness was then determined in the steel thus coarsened. In order to test the uniformity of the material the various specimens were then heated to 1,050° for 30 minutes and air cooled. The hardness then lay between 78 and 82 for No. 4, between 77 and 82 for No. 5, and between 77 and 79 for No. 6. There were indications, though not decisive ones, that the loss of hardness which accompanied the increasing deformation after coarsening at 870° persisted after then refining by heating to 1,050°. Certainly in every case the hardness after deformation followed by heating to 1,050° and air cooling was materially less than after the heating to 1,050° and air cooling prior to the deformation, indicating that the softening effect of deforming plus heating persists even when this heating reaches 1,050°, as is seen on comparing the last two lines.

TABLE 20A.—SOFTENING EFFECT OF GRAIN COARSENESS

Refining initially coarse-grained thermit steel by heating to 1,000° and air cooling softens it.

No.	Brinell hardness under 3,000 kg. for ½ minute			
	Fine	Coarse	Excess of fine	No. of tests
3T	187	177	+10.0	9
4T	104	103	+ 1.0	5
5T	108	99	+ 9.0	6
6T	87	82.5	+ 4.5	4

Note to Table 20A.—Nos. 3T, 4T, and 5T are thermit risers, that is to say, the excess of thermit steel which remains in the riser, which feeds the molten metal to the part to be welded. No. 6T is an ingot of thermit steel made specially for this purpose, and cooled very slowly in its sand mould. The thermit risers have not been analyzed, but it is evident from their hardness and their microstructure that they must contain some carbon or other element. The metal in the coarse state was heated to 800° for 30 minutes while within a closed iron tube set in the electric resistance furnace. The tube, with the specimens in it, was then removed and air cooled. The refining was done by heating part of the same specimen to 1,000° for 30 minutes, and cooling under the conditions used in cooling the coarse material. The heating to 800° was for the purpose of removing the effects of any abnormality in the conditions of cooling from the molten.

TABLE 20B.—SOFTENING EFFECT OF GRAIN COARSENESS

Brinell hardness under various loads of thermit steel in its initial coarse state and after refining by heating to 1,000° for 40 minutes and air cooling.

Group I, a separate spot for each pressure

Line	Pressure in kg.	500	1,000	1,500	2,000	2,500	3,000	3,000 for 5 min.	No. of tests	Average
1	Refined, No. 4T4	97	97	103	106	109	106	108	3	.....
2	Coarse, annealed, No. 4T5	97	101	100	104	106	103	102	3	.....
3	Excess of refined	0	−4	+3	+2	+3	+3	+6	.....	+2.2
4	Refined, No. 5T4	97	99	98	104	110	108	107	3	.....
5	Coarse, No. 5T5	97	99	98	104	108	102	105	3	.....
6	Excess of refined	0	0	0	0	+2	+6	+2	.....	+1.6
7	Refined, No. 6T4	84	80	85	89	87	82	.....	2 or 3	.....
8	Coarse, No. 6T5	74	71	84	88	86	84	.....	2 or 3	.....
9	Excess of refined	+10	+9	+1	+1	+1	−2	.....	.....	+3.3

Group II, the successive impressions made at the same spot

10	Refined, No. 4T4		96	107	100	106	106	91	.....	1	.....
11	Coarse, No. 4T5		89	101	97	105	105	92	.....	1	.....
12	Excess of refined		+7	+6	+3	+1	+1	−1	.....	.....	+3
13	Refined, No. 5T4		94	100	97	96	95	91	.....	5	.....
14	Coarse, No. 5T5		92	99	97	97	96	92	.....	6	.....
15	Excess of refined		+2	+1	0	−1	−1	−1	.....	.....	0
16	Refined	Rolled cold, 10 per cent. reduction, No. 5T6	159	162	164	156	154	150	.....	3	.....
17	Coarse	Rolled cold, 10 per cent. reduction, No. 5T7	150	159	159	150	147	143	.....	3	.....
18	Excess of refined		+9	+3	+5	+6	+7	+7	.....	.....	+ 6.2
19	Refined	Rolled cold, 20 per cent. reduction, No. 5T6	.....	194	181	178	172	.....	.....	2 to 6	.....
20	Coarse	Rolled cold, 20 per cent. reduction, No. 5T7	.....	173	174	166	167	.....	.....	3 to 6	.....
21	Excess of refined		.....	+21	+7	+12	+5	.....	.....	.....	+11.2

Note to Table 20B.—The materials 4T, 5T, and 6T are those described under the same numbers under Table 20A. The thermal treatment is the same as is described under that table.

Group I.—The hardness was determined under varying pressures which were maintained for ½ minute, with the exception of the 3,000 load which in addition was in two cases kept on for 5 minutes. Here each determination was made at a new spot.

Group II.—The general procedure was the same as in group I, except that the successive impressions were made at the same spot. After each impression an interval of 15 or 20 minutes elapsed, and then a new impression with the next higher pressure was made at the same spot, the ball resting in the hole made by the previous pressure.

Lines 13 to 21.—These tests were made on two slabs 4.5 in. × 0.7 in. × 0.300 in. cut lengthwise from the thermit ingot 5T.

For the experiments covered by lines 13 to 21 one slab was heated to 1,000° for 40 minutes for the purpose of refining it. In this heating it was enclosed in a closed iron tube set within the electric resistance furnace. The tube was then drawn and cooled in the air. Then both slabs were heated together to between 750° and 775° for 30 minutes and air cooled within the tube for the purpose of bringing them to the same condition apart from the refining.

The hardness was then determined in both slabs. The average of these results and of other comparable ones made on transverse slabs cut from the same ingot are given in lines 13 to 15. Then these two slabs were reduced 10 per cent. in thickness by rolling cold from a thickness of 0.300 to 0.270 in. They were then heated for 15 minutes in boiling water, cooled, and left for 16 hours. The purpose of the heating was to hasten the elevation of the elastic limit caused by the cold rolling.

The tests indicated in lines 16 to 18 were then made on them. The same slabs were next reduced by rolling to a thickness of 0.240 in. or an additional 10 per cent. of the original thickness. They were then heated to 100° for 15 minutes and subjected to the tests covered by lines 19 to 21 without any intermediate rest.

Attempts made to determine the hardness under a 3,000-kg. pressure led to cracking the specimens.

I have sought direct evidence by tensile tests of Hadfield's 4 per cent. silicon steel in its fine-grained state, and when extraordinarily coarsened by Ruder's[1] modification of Stead's method, heating highly in hydrogen after cold rolling (Table 18); by Brinell hardness tests of ferrite of various degrees

[1] *Trans. Amer. Inst. Min. Eng.*, 1913, vol. 47, p. 569; and *Journ. Industrial and Engineering Chemistry*, 1913, vol. 5, pp. 452 to 458.

of grain coarseness induced by Sauveur's critical overstrain process, with differential straining by tension, by direct compression, and by cold rolling (Table 19); by Brinell hardness tests of thermit steel in its coarse state as cast and after refining by heating to 1,000° (Tables 20A and B); by tensile and hardness tests of very low-carbon steel after various degrees of coarsening by high heating; and by Turner sclerometer tests of the silicon steel. These I will now describe.

490. Direct Tensile Tests of Coarse and Fine Silicon Steel, Table 18.—The sectional grain area of the coarse was about 2,500 times that of the fine-grained silicon steel. This coarsening was accompanied by a loss of 15 per cent. in yield point, 21 per cent. in Turner-sclerometer hardness,[1] and 30 per cent. in tensile strength. That the loss of tensile strength is so much greater than that of yield point, in view of the much greater deformation implied in the determination of the former than in that of the latter, is of the first importance for interpreting the phenomena. That the loss of Turner-sclerometer[2] hardness is intermediate lends support to the hypothesis that the effect of grain coarsening increases with the degree of deformation, for we may hold reasonably that the grain boundary deformation of the fine-grained steel caused by the scratch is intermediate between that caused by the very slight deformation of the whole at the yield point and the great deformation at the time of passing the maximum load.

491. Influence of Crossing a Grain Boundary on the Width of a Turner Scratch.—Such a scratch failed to change width on crossing the grain boundaries of the coarsened silicon steel.

492. Influence of Grain Coarseness Induced by Sauveur's Differential Overstrain on the Brinell Hardness, Table 19.—These experiments followed Sauveur's process of coarsening ferrite by heating it to below *A*3 after a critical degree of plastic deformation.[3] In order to include this critical de-

[1] The Turner sclerometer test consists in measuring microscopically the width of a scratch made by a diamond under a fixed load, and drawn across the face of the specimen under fixed conditions. In the present case the specimens were too thin to be tested fairly by either the Brinell or the Shore hardness test (Turner, *Journ. Iron and Steel Inst.*, 1886, No. 1, p. 175; and "The Metallurgy of Iron," Lippincott, London, 1895, p. 241).

[2] This result does not really conflict with the failure of the Turner scratch to narrow measurably on crossing a grain boundary (§491). In a fine-grained steel the scratch plows across innumerable grain boundaries, of which some reach the initial surface but many do not. The many boundaries thus plowed across in each step of advance of the scratch, acting cumulatively, may well cause the observed 21 per cent. increase of Turner hardness, though any one of them might not suffice to affect this hardness measurably. For that matter the Turner scratch is itself rather irregular, so that the width reported has to be the mean of many measurements. With such irregularity of width, a slight narrowing on crossing a grain boundary might escape notice. That any narrowing which occurred must have been very small is shown by the fact that none was noticeable in any of the three scratches made on the coarse steel. The width of the scratches in the three coarse steels tested was identical, and the same is true of the fine, the difference between coarse and fine as regards the width of the scratch being conspicuous to the naked eye.

[3] Sauveur, *Proc. Internat. Assoc. Testing Materials*, VI Congress, New York, 1912, II, 6. See also Rejto, *idem*, Section A, p. 134; and Chappell, *Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, p. 476. Such heating after Sauveur's critical plastic deformation does not, I find, always lead to coarsening. Thus a tapered bar of the A.R.M. steel was subjected in my tests to a tensile load

gree, a wide range of deformation was given, as described more fully in the note to Table 19, and the hardness resulting from various degrees of deformation was determined, with the results here assembled.

Before interpreting these we should notice two unexpected phenomena:

First, the hardness, after annealing at about 700° or about 850°, decreases progressively as the deformation increases, as shown in each of these six series of Table 19.

Second, though these differences are nearly effaced by later heating to 1,050° and air cooling, yet a certain softening effect of the deformation persists, for the hardness after deformation followed by this 1,050° annealing is less than that after an identical annealing not preceded by such deformation, as is seen on comparing the last two lines of the table.

Because of the first of these laws, the effect of grain size on the hardness appears best as a break in the increase of hardness which accompanies the decrease of the deformation in passing up the several columns of Table 19. This break takes the form of replacing the progressive increase with an absolute decrease on reaching the coarse stage, in cases 1, 4, and 6; the form of a cessation of the increase on reaching the coarse stage in case 5; and the form of an acceleration of the increase on passing up beyond the coarse stage in case 3. In case 2 the variations in grain size are too small to cause a measurable break in the hardness curve.

When we consider that a 2,500-fold increase of the grain size in the silicon steel of Table 18 caused a loss of only 15 per cent. in the yield point, we need hardly wonder that the break in the hardness curves disclosed by the relatively slight variations in grain size of Table 19, from 3- to about 20-fold, are so small as not to exceed greatly the errors of measurement. All that we can say is that in every case in which there is a marked change in grain size, the break which occurs agrees in sign with the hypothesis that grain coarseness has a slight softening effect.

493. Refining Coarse Grain by Heat Increases the Hardness.—Three ingots of thermit steel, $4T$, $5T$, and $6T$, of which the last was nearly pure ferrite, had their hardness determined both in the coarse-grained state in which they originally solidified and after grain refining by heating to 1,000°, and under pressures varying from 500 to 3,000 kg., with the results shown in Tables 20A and B. In order to efface the effects of any abnormality in the rate of cooling from the molten, the coarse specimens were heated to 800° and air cooled under conditions like those of the grain-refined specimens.

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which varied continuously from 40,640 to 49,000 lb. per square inch, over a length of 4 in. Under this stress, which at the smallest section was greater than that which broke a companion test piece cut from the same billet, the specimen necked greatly, so that its section was reduced by 21.4 per cent. Hence some part of this tapered length should have undergone the critical deformation; yet heating the bar for 22 hours at 680° failed to coarsen any part of it noticeably. Professor Sauveur finds that this coarsening takes place most readily when the carbon content is between 0.05 and 0.07 per cent., and that in order that decided coarsening shall occur the carbon content should be between 0.04 and 0.12 per cent. (private communication, Aug. 21, 1915). (See also note to §585.)

In group 1 each impression was made at a new spot, whereas in group 2 six successive impressions were made on each spot with progressively increasing loads, at 15- or 20-minute intervals, the ball in each test resting in the depression made by the preceding test. For the tests represented by lines 1 to 15 no further preparation was made, but for those represented by lines 16 to 21 the metal was rolled cold, in order that the effect of this cold deformation on the relative hardness of coarse- and fine-grained metal might be learned. Between the cold rolling and testing the specimens were heated to 100°, in order to develop quickly the effects of the cold rolling, which in the cold develop only slowly.

The results, summarized in the last column, are striking in showing that whereas the hardness of the fine exceeds that of the coarse, on an average of each set, by at most 3.3 Brinell numbers, or 2 numbers on an average of the whole, when there has been no prior deformation, this excess rises to 6 numbers when there has been a prior 10 per cent. reduction by cold rolling, and to 11.2 numbers when this reduction has been 20 per cent.

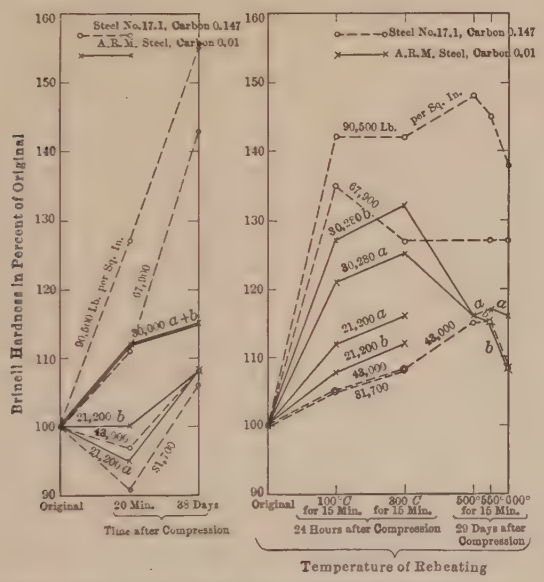

FIG. 78.—Influence of rest and heating after compression on the Brinell hardness.

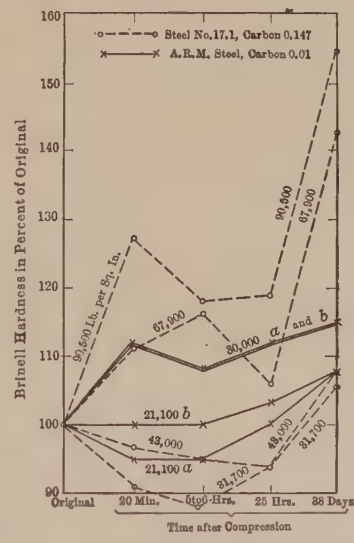

FIG. 79.—Changes in the hardness shortly after the deformation.

It might be expected that to make the successive impressions in the same spot would lead to an increase of the apparent hardness, because the hardening change incited by each impression would continue during the interval before the next impression, quite as I had found that retarding the Brinell test increases the apparent hardness. But instead of this effect the opposite is suggested. Thus comparing lines 1 and 2 with 10 and 11, and 4 and 5 with 13 and 14, the increase in hardness from the 500- to the 3,000-kg. impression is greater when each impression is made on a new spot than when all are made on one spot. This may be referred to the slowness with which the hardening caused by slight deformation sets

in, as shown in Figs. 78 and 79. There we see that even after 5 to 6 hours' rest there was often an appreciable loss of hardness, which was recovered later. Yet under greater deformation the increase of hardness was measurable even at the end of 20 minutes.

494. Grain Coarsening by Overheating Weakens More than it Softens.—My preliminary experiments, while confirming the general observation that long high heating, say to 1,300° or higher, lessens the tensile strength materially even in the case of nearly pure ferrite, yet show that it affects the hardness very little.

495. Discussion of the Foregoing Evidence. The Boundary Resistance, though Very Slight at First, Increases as Deformation Proceeds.—The slightness of the boundary resistance prior to deformation is shown in Table 18 by the slightness, only 15 per cent., of the lowering of the yield point even by the 2,500-fold increase of grain size in the silicon steel; by the failure of the Turner scratch to narrow on crossing the grain boundaries of the coarse silicon steel; by the slightness of the break in the hardness-deformation curve of the Sauveur differentially strained specimens, Table 19; and by the slightness of the excess of the hardness of the refined over that of the coarse unstrained thermit specimens of Table 20B.

It tallies with this that the presence of Neumann lamellæ does not affect the hardness of ferrite measurably, as shown by Stead (§606). For even if the slip planes on either side of the contact of a twinned lamella with the enclosing metal should register, the slip itself must in general change direction here, as is shown by the zigzagging of the slip bands on crossing a twinned area. This thrusting of a slipping stratum from the enclosing metal into the twin lamella, and the abrupt change of direction of the slip at the contact plane, should cause a break-up of some of the crystalline metal, exactly as in the case discussed in §481.

That the boundary resistance increases as the deformation proceeds is indicated concurrently by the slightness of any initial boundary strength, as shown by the failure of the Turner sclerometer scratch to narrow on passing a grain boundary; by the progressive increase in the effect of the grain coarseness of the silicon steel as the deformation increases from that of the yield point to that of the Turner sclerometer test, and thence to that of the tensile strength or maximum load, in Table 18; by the three- to six-fold excess of the effect of grain size on the hardness in the previously cold rolled specimens of Table 20B over its effect in those not previously cold rolled, an excess, moreover, which increases with the degree of that cold rolling; and by the greater effect of grain coarsening caused by overheating on the tensile strength than on the Brinell hardness.

Further evidence in this same direction is that, though in its first stages deformation may occur close to if not truly in the grain boundaries (§407), the final stages, as shown by the path of rupture, not only avoid those boundaries but actually seek the grain centers (§694).

496. Further Evidence.—It is in harmony with this hypothesis of a

progressive increase in boundary strength as deformation progresses that the approach of the slip bands to the terminal grain boundaries becomes less regular as deformation proceeds, as it naturally would if the resistance of those boundaries to the propagation of slip across them increased with the progress of deformation. But this evidence is not of real weight for our present purpose, because this progressive increase of irregularity with increasing deformation is readily explicable as the result of the accumulation of amorphous metal along the slip planes, sheathing the crystalline blocks which slip past each other (§677).

497. The Way in which Slip Bands Originate and Propagate is Consistent with the Theory that Disregistry Opposes Slip.—The starting of the slip bands in part at the grain centers and in part at the grain boundaries, as is shown in §414 and Fig. 60, is consistent with the hypothesis that the least resistance to slip is midway between terminal boundaries, and that slip may either originate within a given grain and about midway between terminal boundaries, spreading thence toward one or both of them, or may originate at terminal boundaries because the existing stress is localized here, and may propagate thence first and most readily to about midway toward the opposite terminal boundary, and later fully to that boundary, as the stress increases. For in any event the stress must reach any given grain through its boundaries, and here the stress is likely to be most sharply localized. This localizing of the stress at the end of a slip plane may outweigh the end-support which the slip plane receives there because of disregistry with the adjoining grains, with the result that the slip starts at the boundary and spreads thence inward; or it may not outweigh that support, with the result that the slip starts at about the middle of the length of the slip plane, where its end-support is least effective, and spreads thence toward the terminal boundaries.

The very origin of slip may well be at a grain center where the end-support is least, yet the propagation of that slip from that grain into its neighbor naturally starts in that neighbor from its boundary, and thus causes a slip band which too enters that grain from its boundary. Thus it may well be that the slip in grain I of Fig. D of Plate 23 has induced the slip in grain II and thence in grains III and IV, and that in each of these the slip band started at the left-hand boundary of the grain and spread thence in the direction of the arrows. So the short slip bands at *a* in grain II of Fig. A of Plate 23, and in the various grains of Figs. B and C of Plate 25, may have been induced from without the grain in which they occur. Deformation passing from right to left in grain III of Fig. A of Plate 23 might be within the elastic limit in that grain, and yet create a stress at the edge of grain II which would pass the elastic limit of the slip planes along which the slip bands at *a* have formed. Or, if beyond the elastic limit in grain III and hence causing slip, the slipping plane may be so nearly normal to the plane of the paper that the component of the step parallel to the paper is invisibly narrow. It is, of course, this component only that ever can be seen.

498. Why the Slip Bands Curve.—Two reasons appear why the slip bands, even if they are in essence strictly geometrical, may yet in appearance be very ungeometrical and irregular: First, each may be an irregular integration of innumerable short geometrical fractions, and second, each may be deformed subsequent to its formation. Let us consider these.

499. Irregularity May be Caused by Alternations between Conjugate Slip Planes.—Each curved slip band may in fact be composed of two conjugate sets of straight bands, inclined to each other and following two conjugate sets of slip planes. After following one of these directions for a very short distance, slip turns sharply into the other, and by means of it steps over to a new slip plane parallel to the first, thus changing from one direction to the other by unrecognizably short steps so as to simulate a curve, somewhat as a ring made up of a great enough number of fine enough sawteeth[1] would look like a true circle.

The fact that these steps are sometimes recognizably large makes this explanation the more credible. Note how in Fig. A of Plate 24 the slip bands follow a common direction in the odd-numbered areas, 1, 3, and 5, and a different but still common direction in the even-numbered ones. It is possible to refer this case, and indeed any case in which the individual paths along the several slip planes of two conjugate sets are detectably long, to multiple twinning.

A difficulty with the conjugate slip theory is that cases in which such detectably long straight lines integrate into irregular slip bands are not common. On the other hand, if these deviations from straightness represented fluid motion, they ought to be smooth, as for instance in grain IV of Fig. C of Plate 23, and at *ab* in Fig. B of Plate 31. Their usual irregularity, which is often very pronounced, points rather to this present explanation of pathlets of irregular and short length along conjugate planes.

For that matter, many of the lines in Fig. B of Plate 35, which at first sight look curved, on closer examination are seen to be made up in large part of short straight parts.

The conjugate slip hypothesis tallies with the impression which one receives that, after allowing for difference caused by the degree of deformation, the slip bands are more regular in coarse than in fine-grained steel, as if the development of the crystalline organization which accompanies the grain coarsening increased the continuity or solidarity of the individual slip planes. Note for instance the straightness of the slip bands in the coarse grain of Fig. D of Plate 22, in Fig. B of Plate 24, of the A.R.M. steel coarsened by long high heating, and in Fig. D of Plate 33 of the coarsened silicon steel. Should this impression be confirmed, it might give us an additional means of detecting overheating.

Indeed the general elaboration which long high heating gives the crystalline organization would in itself tend to this continuity of slip along a given

[1] For evidence to this effect see Rosenhain, *Journ. Iron and Steel Inst.*, 1904, No. I, vol. 65, p. 343, and Figs. 6 to 9, pp. 344–5.

plane, for one feature of this elaboration would naturally be to cause the difference in strength between the slip planes and random planes to increase, and thereby to approach the limit toward which it tends. This increase in itself would increase the tendency to slip to remain in a given slip plane in which it has started, instead of stepping out of it through a path of greater resistance. This is only one of the consequences of this elaboration. I have elsewhere traced other natural consequences of it.[1] Among other things, the long high heating, with its accompanying mobility, would assist the expulsion of foreign matter to the grain boundaries, leaving so much the less within the grains to interfere with the elaboration of their crystalline organization.

500. The Genetic Rigidity of Ferrite Helps Explain the Irregularity of Its Slip Bands.—The fact that the slip bands in ferrite are, as a whole, less regular than those of other metals tallies with this conjugate slip hypothesis, for whereas the crystalline structure of most metals forms under the favorable environment of quiescent solidification from the molten state, that of ferrite forms under the extremely unfavorable conditions of the transformation of the firm and stiffly viscous austenite in cooling through the transformation range from about 900° to about 700°, conditions which clearly cannot give such freedom of movement to organize in strict obedience to the crystalline impulses as there is when, in solidifying, one particle after another turns in the remaining molten mother metal so as to conform with the already solid ranks to which it now attaches itself. Here only the slight friction of the surrounding molten particles impedes strict obedience to the crystalline impulses, whereas in the transformation of the solid austenite into ferrite, each new particle which transforms and attempts to conform with the already transformed and assembled ferrite is hampered by the obstinate viscosity of the still untransformed austenite sticking to its other sides.[2]

The behavior of lead accords precisely with these ideas. In an ingot which has solidified tranquilly the slip bands are usually straight, but if this favorable organization out of the molten is replaced by reorganization in the solid state, they are crooked like those in ferrite. This occurs, for instance, when lead is first greatly deformed so as to inequiaxe its crystals, then annealed so as to allow a new and equiaxed structure to reorganize out of the ruins of the old, then repolished and restrained.[3]

501. The Deformation of Slip Bands by Later Slip along Conjugate Planes.—The very straight slip bands of Fig. E of Plate 45 are crossed by two

[1] "Life History of Network and Ferrite Grains," *Proc. Amer. Soc. Testing Materials*, 1911, vol. 11, p. 297. The symptoms which I there pointed out of the elaboration of the crystalline organization by long high heating were: (1) its opposing the transformation in cooling; (2) its increasing the stability of the red-hardness of high-speed steel; (3) its retarding the coalescence of the internal ferrite; (4) its increasing the perfection with which the austenite assembles the pro-eutectoid element into its own cleavages; (5) its making the cellular structure more persistent; and (6) its retarding the coagulation of sorbite into pearlite.

[2] Compare Ewing, *Journ. Iron and Steel Inst.*, 1904, No. I, vol. 65, p. 372.

[3] See Rosenhain, *Journ. Iron and Steel Inst.*, 1904, No. I, vol. 65, pp. 342, 346, and Figs. 10 and 11.

other sets at angles of about 60°. On carrying the deformation a little further, this straightness changed to a marked reversed curvature, and simultaneously the curvature of the twin of Fig. AA of Plate 30, from another field in the same specimen and at the stage of Fig. D of Plate 45, was exaggerated as shown in Fig. AB of that plate.

In this specific case the curvature which later slip along a conjugate set of planes creates in the earlier set of slip bands, and in the twins of Figs. AA and AB of Plate 30, happens to be regular, but it need not in other cases. If the faulting is absolutely regular both as to distance between slips and as to throw at each slip, it will not disturb the initial straightness of a slip band. A regular deviation of either the throw or the frequency of faulting will change initial straightness to a smooth curve, as in Figs. AA and AB of Plate 30, in which the slips are so close together and their throw so small that the individual faults cannot be detected on this magnification. But any irregularity of either the frequency or the throw of the faulting, or of both, will lead to correspondingly irregular deformation of the initially straight slip band or twin. Such irregularities of faulting are not antecedently improbable, and thus are reasonable explanations of the irregularities of the slip bands.

In short, while the straight parts of the slip bands, and their following definite crystallographic directions, are reasonably referred to crystalline motion, they do not show whether that motion is by slip or by twinning; and the curvatures and irregularities are wholly equivocal, for though they might represent fluid motion, yet they are compatible with exclusively crystalline motion, because they might be caused either by frequent irregular stepping across from one to another set of conjugate planes, and thus be congenital, or by the irregular faulting by a later set of slips.

This is equally true of the irregularities and roundings of the profiles of slip bands as seen in cross-section, and of the rounded and sausage look shown in Fig. G of Plate 22. Still another cause of irregularity is suggested by Fig. 75, p. 345. If the undercutting should be still greater than is here shown, the edge or nose of the protruding step might be so extremely thin that it would offer little resistance to being stretched out by its adhesion to the superficial layer of amorphous metal caused by the polishing which habitually precedes our observation of slip bands. Hence may arise irregular outstretchings of the edges of these steps. Note that in Fig. C of Plate 22, which suggests that the slip planes are very flat, the irregularity of the slip bands is extreme.

We shall see in §677 that the sheathing of each crystalline blocklet with Beilby's amorphous metal on deformation gives a way in which the strictly crystalline motion of the crystalline part of the metal may yet be accompanied by fluid motion within the amorphous part.

502. An additional reason why the slip bands in ferrite curve, offered by Heyn,[1] is that their formation is accompanied by fluid movement. This

[1] Martens-Heyn, "Materialenkunde," II, A, Springer, Berlin, 1912, p. 223.

may be considered in connection with the evidence summed up in Chapter 27, bearing on the existence and extent of fluid movement in general in metals. But incidentally one hardly sees why fluid movement should occur rather in ferrite than in other metals, and rather in metal which has reorganized in the solid state than in that which has organized out of the molten state. The frequent presence of foreign bodies, cementite, slag, etc., in ferrite might well tend to divert deformation from the crystalline to the fluid form.

503. Why the Slip Bands Change Direction at their Ends.—This may be referred to disregistry (§479). Thus in Fig. 65, p. 337, the descent of row 3 compels a descent of the cubes against which it impinges in the grain *MKLN* below it, but in a different direction. If slip were exactly as easy in the direction of arrow *r* as in that of *r'*, both directions might be followed, so that the resultant direction of displacement might be the same in the lower as in the upper grain, and two sets of slip bands would be set up. But such exact balance is not to be expected, and one direction, for instance *r*, is likely to be easier than *r'*, as is proved by the habitual starting of slip along a single set of roughly parallel slip bands in any one grain. In short, on reaching the terminal boundary *KL* the direction of slip swings to the left, from the vertical, *tt'*, to *r*.

504. The Analogy of Liquids.—If it were not a set of crystalline grains but a liquid that thus changed its direction, for instance at the elbow *KL* of Fig. 80, it would begin swinging round to the left before it reached *KL*, so that the path of any given particle below *KL* would be a smooth continuation of its path above.

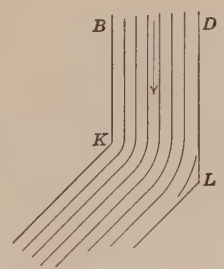

FIG. 80.—Flow accommodates itself to a sharp bend.

It is natural that the power which slip has of curving, whether through conjugate slip or through fluid motion, should be exerted in such a way as to facilitate the propagation of the slip from any one grain into the next grain, by accommodating itself to the intergranular disregistry. It is easy to believe that such facilitation is offered by just such curvature as we notice, like that of the liquid at the elbow *KL* of Fig. 80.

Thus in Fig. D of Plate 23 the slip is continuous from grain I to grain IV. Not that any single slip band can be traced with confidence across all three boundaries, but that the slip bands in these four grains form a set, one band at least being traceable across each boundary. It is natural that this set should be a link in a series which forms the path of least resistance to the existing stress across the four grains collectively. To fix our ideas we may assume that the outcrop of this set least favorably disposed to resist slip, and hence most favorably disposed to slip, is in each grain parallel to the arrow there sketched. It is easy to believe that, taking these four grains as a whole, the easiest slip is exactly what we find, running parallel to the arrow throughout the greater part of each grain, but curving round as it approaches the grain boundary in such a way as to connect smoothly

with the arrow direction in the next grain, by means of a continuation of the same curve on the other side of the boundary. In this way the shape of the curve comes to be like that of a liquid in passing a series of elbows.

So with the curvature in passing from grain I to II and II to IV of Fig. C of Plate 23, and from I to IV of Fig. E of Plate 23. In this last case the dominant direction in grain IV is nearly 90° away from that in grain II, yet the two are brought into a single system by the long sweep of the slip bands, for instance just at the left of the numerals IV.

There may be cases in which we cannot explain with confidence why a certain specific curvature exists in the approach to a given grain boundary. But this is to be expected. The problem is a complex one in descriptive geometry, and many of the data are necessarily hidden, such as the inclination of the various slip planes to the plane of the paper and hence to each other, their relative intrinsic strength, and the strength of the component of the stress parallel to them.

505. Resolution of the Slip Steps into Permanent Elongation and Contraction of Area.—If in Fig. 76, p. 346, the slip is parallel to the pull, and if we resolve any one of these little steps such as aC' into two components, one vertical and one horizontal, the horizontal component is a minute fraction of the total permanent elongation of the test piece, and the vertical one is such a fraction of the contraction of area. Hence on looking down on such a polished surface after deformation, the total width of the black bands measured in the direction of the pull should represent the total permanent elongation.

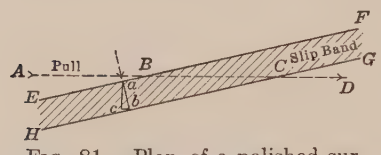

FIG. 81.—Plan of a polished surface on which a slip band has formed oblique to the pull AD.

But if, as in Fig. 81, the direction of pull is AD and the slip is in the direction ab, then the component of this slip parallel to AD is only cb, a small fraction of the intercept BC which the step would blacken, and only a fraction of the width, ab, of the black band. Hence the blackening of any surface by the slip bands, while it cannot be less than the permanent elongation of the specimen which that slip causes, yet may be very much greater, and hence, unless interpreted cautiously, may give an exaggerated impression of the actual deformation.

CHAPTER 20

BEILBY'S AMORPHOUS THEORY

506. Introduction.—This chapter takes up Beilby's amorphous theory, the first to offer a reasonable explanation of the mechanism of plastic deformation in metals, and in particular to explain ductility, the retention of continuity during the deformation of these crystalline masses, and the extraordinary fact that their tensile strength is usually greater than their elastic limit.

507. The First Important Slip Ought to Precipitate Rupture.—A stress great enough to force the particles aa', bb', etc., in Fig. 82, facing and holding fast to each other across a given crystallographic plane AB, to let go their hold and slip past each other would, if continued, naturally be expected to cause rupture, for the resistance to shearing which the cohesion of these particles offers certainly ought to be greater before than after they have begun to slip. Yet in the tensile test, though slip begins before the stress reaches the yield point, the stress needed to cause rupture, that is the tensile strength, may be nearly twice the yield point stress absolutely, and far more than twice as great per unit of section as this is reduced by the stretching between the yield point and the maximum load. Here we have the paradox that the cohesion between the particles thus facing each other across AB seems to increase progressively and greatly as they go slipping past each other, which is as if the friction of motion were greater than that of rest.

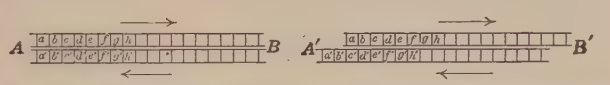

FIG. 82.—During slip the crystal units change partners.

The very first slip in any one grain, of course, need not lead to rupture, because it may well occur along a crystallographic plane very unfavorably disposed, and the more favorably disposed grains at the terminal boundaries of that plane may be able to arrest that slip without passing their elastic limit. This same principle might explain the arrest of slip after it had extended to several grains, yet it cannot explain why, after slip has occurred along many different slip planes in every grain, the test piece not only does not slip forthwith into fragments on the mere continuation of the existing stress, but actually refuses to stretch further till the stress is increased, and, in spite of the ever decreasing size of its cross-section, persistently opposes greater and greater absolute resistance to every further increment of stretch. Still less can it explain why, if the test is interrupted, a still greater stress will be required after rest to cause further stretch, proving that the cohesion has increased during rest.

508. Beilby's Amorphous Theory of Crystalline Slip.—Beilby's explanation[1] of these phenomena is that, just as two strata of rock slipping past each other during an earthquake in causing the familiar slickensides or polished contact pulverize the particles which they thus rub off, so when the slip occurs along the cleavage plane AB of Fig. 82, minute particles must needs be torn away from the contact faces of aa', bb', etc.; and that these particles are of a lower order of magnitude than any possible crystal unit, and hence are necessarily amorphous, that is, free from crystalline organization.

Further that, whether because the small quantity of heat liberated by the friction is so closely localized that it actually melts these particles quite as the close localization of the heat generated by the friction of an earthquake slip may melt the rock locally and thus lead to a lava flow, or for some other reason, the metal torn away by the slip is very mobile momentarily.

Finally, that, whether because the extremely thin layer thus made mobile cools too rapidly to be able to resume the crystalline state, or because it is of less than crystal unit thickness and hence cannot recrystallize, or for whatever other reason, the metal passes from the mobile into a solid noncrystalline and hence amorphous state, which is stronger than the normal crystalline state. This greater strength is natural, because the amorphous state lacks the cleavages and slip planes which weaken crystalline substances. Hence the strengthening and hardening effect of every kind of plastic deformation, which is identical with "overstrain," "interstrain" and "cold work."

Thus we imagine three distinct states, first the initial crystalline state, next a temporary very mobile state, leading rapidly to the final solid amorphous state. The thickness of the amorphous layer caused by a single slip is likely to be extremely thin, perhaps about 20 to 30 $\mu\mu$ thick.[2]

[1] Beilby, "The Hard and Soft States in Metals," May Lecture, *Journ. Inst. Metals*, 1911, No. 2, vol. 6, p. 5. Also "Surface Flow in Crystalline Solids under Mechanical Disturbance," *Proc. Royal Soc.*, 1904, A, vol. 72, p. 218. "The Effects of Heat and of Solvents on Thin Films of Metal," *idem*, 1904, A, vol. 72, p. 226. "The Hard and Soft States in Ductile Metals," 1907, A, vol. 79, p. 463. "The Influence of Phase Changes on the Tenacity of Ductile Metals at the Ordinary Temperature and at the Boiling Point of Liquid Air," *idem*, 1905, A, vol. 76, p. 462. "Surface Flow in Calcite," *idem*, 1909, A, vol. 82, p. 599, "Transparence or Translucence of the Surface Film Produced in Polishing Metals," 1914, A, vol. 89, p. 593. "The Hard and Soft States in Metals," *Philosophical Magazine*, Aug., 1904, p. 257 (from the Faraday Soc., June 9, 1904).

"The theory of flow and resolidification in a vitreous-amorphous state for which I am responsible . . . postulates that layers of molecules many molecules in thickness have the mobility of the liquid state conferred on them for a brief period; that it is the transient existence of this mobile phase which makes slip and movement among the lamellæ possible and easy, and that it is the sudden resolidification of these mobile layers into a noncrystalline or vitreous condition which arrests deformation under a given deforming stress. The original surfaces of any slip are now cemented together by the more rigid material and new surfaces of slip are only developed by higher stresses. The plasticity of the crystalline state is thus gradually used up and the aggregate as a whole becomes more and more rigid. When this has reached a certain stage further increase of stress leads, not to plastic flow but to disruption." (G. T. Beilby, private communication, Mar. 26, 1915.)

Rosenhain, "Metals, Crystalline and Amorphous," *Engineering*, 1913, vol. 96, p. 509, also "Physical Metallurgy," 1914, p. 246, Van Nostrand Co., New York.

[2] G. T. Beilby, private communication, Mar. 26, 1915. 1 $\mu\mu$ = 0.000,001 mm.

This mobility is necessary to explain the retention of continuity during slip. Yet it implies great weakness, which, if it remained for any appreciable length of time, would weaken for instance a tensile test piece as a whole greatly, and thus prevent completely the actual increase of strength which occurs during deformation. To explain this continuous increase of strength during the alleged amorphization, we must needs assume that the passage from the initial strong crystalline state, through the weak mobile state, to the still stronger final amorphous state, is so rapid that at any given instant only a few out of the total number of planes along which slip has occurred or is occurring are lined with mobile metal, the metal along any one plane passing through and beyond the mobile to the solid amorphous state before the metal along any great number of other planes has become mobile.

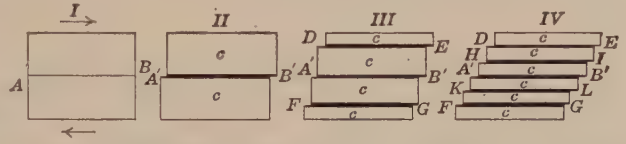

FIG. 83.—Progressive depletion of the crystalline metal during slip, on Beilby's amorphous theory.

Though the metal thus quickly becomes solid and strong, yet it may continue to increase in strength for a long period thereafter, though at an ever decreasing rate, somewhat as cement does.

Because slip along any one plane changes the metal which lines that plane to the amorphous state, as slip proceeds the proportion of amorphous metal must increase and that of crystalline metal capable of passing through the mobile to the amorphous state and thereby of permitting plastic deformation must decrease, somewhat as sketched in the successive stages I to IV of Fig. 83. At stage II the first slipping plane *AB* has become lined with amorphous metal; at stage III three planes have thus become lined, *FG*, *A'B'* and *DE*, the residual crystalline metal *ccc* having decreased in quantity correspondingly; and at stage IV five planes, *FG*, *KL*, *A'B'*, *HI*, and *DE* are lined thus, with further depletion of the mobilizable crystalline metal.

509. The Essence of Plasticity.—In this view plasticity is the ability which some crystalline substances, metals and others, have of passing rapidly through a weak mobile phase and reaching very quickly a new and strong phase. This new phase is believed to be amorphous, and to lack the power of entering into the mobile phase until it has been returned to the crystalline phase, for instance by heating. Hence the amorphous phase is by assumption wholly unplastic, that is to say perfectly brittle.

Because it is brittle it must break up into fragments during each successive stage of the deformation, as it is forcibly deformed beyond its elastic limit by the deformation of the enclosing mass. The cracks in it which thus form in any one stage of the deformation are healed by the inflow of the mobile phase which forms simultaneously.

510. The Exhaustion of Plasticity.—Clearly this must come to an end when the remaining healing mobilizable crystalline metal bears so small

a ratio to the always cracking brittle amorphous metal that it no longer suffices to heal these ever forming cracks completely enough to prevent their propagating across the specimen, and thus causing rupture.

511. Annealing, by giving mobility for an appreciable time, permits the return of the amorphous metal to the crystalline state and thus the renewal of the power of accommodation to plastic deformation by passing, one leaflet at a time, through the mobile to the amorphous state. Thus it is that annealing restores completely the initial ductility and most of the other initial properties.

512. Analogous Cases of Amorphousness.—That this supposed quick passage through the alleged mobile state should result not in a crystalline but in an amorphous state, tallies with the familiar observation that glass, many slags, and transparent candy, when cooled relatively rapidly, retain the amorphousness of the molten state, yet if cooled slowly enough become crystalline, that is devitrify, and simultaneously become softer and less transparent or translucent. In making glass, candy, and varnish our care is directed toward preventing devitrification. Strongly as water tends to crystallize in freezing, it too remains amorphous if solidified fast enough.

The most important parallel case of amorphousness is that of the polished surface of metals and of other crystalline substances in general. A variety of evidence shows that polishing has this specific effect of making the polished surface amorphous. For instance, the scratches made on the surface of even the brittle and crystalline antimony by rough polishing can be completely flowed over by fine polishing, covering the surface as with a thin varnish wholly free from crystalline structure. On appropriate etching this amorphous antimony varnish can be removed, disclosing the old scratches over which it had been spread. In copper and probably other metals this metallic varnish is transparent or translucent, so that cavities beneath can be seen through it.[1]

An amorphous but very thin layer of metal is conceived by Rosenhain and others to exist in the boundaries between the adjacent grains of the metals in general, iron included, as explained in §696, essentially because the gusset spaces between the disregistering crystalline blocks of the differently oriented grains cannot be completely filled by crystalline blocks, but must needs leave minute spaces which, because smaller than crystal units, can be filled only by amorphous metal.

513. The Changes which Plastic Deformation Causes in the Physical Properties are Those which Amorphization should Cause. The Hardness.—The slip planes of the crystalline state are evidently sources of weakness, of low cohesion. In that passing from the crystalline to the amorphous state necessarily effaces these slip planes and all other planes of low cohesion and thereby strengthens the mass, it ought to increase the hardness, which in the last analysis is only one measure of cohesion.

[1] Beilby, *Proc. Royal Soc.*, 1904, A, vol. 72, p. 218.

514. The Tensile Properties.—Cold plastic deformation of all kinds, for instance by wire-drawing, at first raises the elastic limit as well as the tenacity, but if pushed beyond a certain stage in turn lowers both, while it lessens the ductility continuously, whether measured by the elongation or contraction of area. These are exactly the results to be expected from the progressive substitution of the hard, strong, brittle amorphous iron for the initial crystalline iron in the way described.

To fix our ideas let us consider the case of wire-drawing, remembering that the increase and later decrease of strength caused by it, and now to be explained, are determined by a tensile test which itself proceeds in part through plastic deformation.

That the loss of ductility with the increase of brittle amorphous metal should be continous is evident.

That the elastic limit per unit of existing section should increase in wire-drawing with the increase in the proportion of the stronger amorphous metal is natural enough. That it should later reach a maximum and then decrease when the deformation is pushed too far, is equally natural. This maximum may represent the stage in the progressive accumulation of amorphous metal at which this strengthening effect is in turn outweighed by the growth of internal cracks, caused by the depletion of the crystalline metal by the sum of the deformations, first in the wire-drawing itself, and second in the tensile test by means of which the influence of the wire-drawing on the elastic limit is determined.

Or the maximum tenacity and elastic limit may simply represent that proportion of amorphous to crystalline metal in which the peculiarities of each react to the best advantage on those of the other, a proportion as impossible to predict as the most hygienic ratio of play to work.

515. The increase of volume, the lightening, caused by plastic deformation is a natural result of amorphization, for the random packing of the amorphous state would naturally leave greater intermolecular voids than would exist in the regular crystalline massing. Builders express this roughly by saying that if individual stones exactly equivalent to those of the walls of a house were thrown in pell mell they would fill the house.

The loss of density caused by tensile deformation might be explained mechanically, but that caused by compressive deformation cannot readily. That the lightening is not due to the formation of voids is shown by the almost exact return to the initial density on annealing.

516. The Heat of Solution.[1]—That plastic deformation increases the heat of solution is what should follow from amorphization. When crystalline metal is dissolved, as much heat is absorbed in breaking up its crystalline structure as was evolved in the act of crystallization. When amorphous metal is dissolved the total heat evolution is greater because this absorption

[1] Beilby, *Journ. Inst. Metals*, 1911, No. 2, vol. 6, p. 22. Berthelot referred the marked differences as regards heat of solution between hammered, annealed, and electrolytic silver to allotropic changes.

is absent. The amorphization of part of the metal by plastic deformation therefore should lead to a proportional increase in the actual heat evolution on later dissolving the metal.

517. The increase in the solution pressure, as shown by the faster etching of plastically deformed than of annealed metal and by the familiar tendency of steel to rust locally where it has been indented[1] or otherwise deformed, whether before or after its last polishing, may be explained by three coöperating causes. The plastically deformed metal etches faster because being lighter, that is bulkier and less closely packed, it offers greater surface for attack; because its amorphous metal lacks the crystalline bond which in itself opposes solution and every other kind of attack; and because in being a mechanical mixture of amorphous and crystalline metal, it has local differences of potential, which are such frequent accelerators of corrosion.

518. The translucency which polishing gives metallic surfaces is not only in general harmony with the familiar observation that the devitrification, that is, the crystallization, of initially amorphous glass, slag, and candy lessens or destroys their transparency, but a natural consequence of the amorphousness. For among the most striking properties of the crystalline state are its specific influences on light, and it is but natural that among these influences should be the extinction of light, that is its conversion into other forms of energy.

The loss of electric conductivity on plastic deformation needs further study. One may imagine a certain baffling effect from the innumerable internal surfaces separating crystalline from amorphous regions.

519. The Amorphous State and the Rise from the Yield Point in the Stress Strain Diagram.—Reserving a fuller discussion of this for another volume, we may here recognize certain features. The arrival at the yield point in the tensile test means that the stress has now reached an intensity which causes slip to occur simultaneously along a very great number of scattered internal planes. That the test piece, in spite of the loss of sectional area by stretch, soon stops stretching if the load is now held constant, means that this loss of area is more than offset by the strengthening due to the progressive setting of the mobile metal, generated by the slip and permitting the slip to occur without rupture, into rigid and stronger amorphous metal. The continued rise of the stress strain diagram thence to the maximum load means that the continuing loss of sectional area is progressively outweighed by the continuing generation of stronger amorphous metal through the slip itself. The gain of strength during rest means the slow setting of amorphous metal, comparable with the slow setting of amorphous glue.

Sections 629 to 630 show that, though great deformation increases the hard-

[1] See Brearley, "The Use of Microscopic Methods," Sheffield Soc. of Engineers and Metallurgists, Nov., 1909–1910. Also Fremont, *Bulletin Soc. d'Encouragement*, 1903, Part 2, p. 359, who shows interestingly how, on prickpunching a pattern on a steel surface, grinding away the immediate surface, polishing, and straining, the pattern reappears, in this case through difference not in solution pressure but in hardness.

ness at once, slight deformation causes a temporary softening, followed during rest by hardening. Interpreting hardness and strength as closely related manifestations of cohesion, we may surmise that the drop of the beam of the tensile test represents that relatively slight deformation which softens, and that the arrest of the drop and the renewal of the rise of the diagram represent degrees of deformation so great as to be strengthening immediately.

From these considerations I became convinced that careful measurements would show that the greater time available for this setting due to retarding the tensile test must needs result in greater tensile strength, and that the current assertions to the contrary must represent faulty measurements. Soon after documents were circulated supporting my inference.

520. The effacement of the effects of plastic deformation by heating agrees with the theory that deformation amorphizes. For such heating not only causes in general the passage from the metastable to the stable state, but in particular causes the passage from the amorphous to the crystalline state, the amorphous being the metastable and the crystalline the stable.

Thus heating the undercooled and amorphous solids, glass, fused quartz, clear sugar candy, sulphur, and like amorphous bodies devitrifies them, *i.e.*, changes them to the crystalline state, relaxing the molecular inertia which the cold causes and thus enabling the molecules to obey their impulse to group themselves in crystalline order. Tempering hardened steel by reheating acts through enabling the molecules to abandon the undercooled solid-solution state of gamma or beta iron and obey their impulse to group themselves as ferrite and cementite.

521. Slip is not Confined to Nominally Plastic Metals.—The bending of the bar of ice sketched in Fig. 62 is only a single case in point. Crystals of apparently perfectly brittle minerals, kyanite for instance, are often found imbedded in rocks and bent in a way which indicates that they too have been squeezed by slip into a shape very different from that in which they were first formed, and without rupture.[1] In these cases the pressure is on all sides, a condition favorable to the retention of continuity, as is familiar to the metallurgist from the facts that the edges of plates crack in rolling, though the upper and lower surfaces remain intact, the plate being in effect compressed on all sides in all parts except at its edges; and that the deformation which the metal can undergo without rupture is greater in supporting operations like punching and wire drawing than in tensile testing. The forcing of the particles toward each other during deformation naturally combats their tendency to part from each other and crack apart, as they do when the deformation is accompanied by a pull instead of an all-pervading push.

Indeed even an apparently perfectly brittle metal like antimony can be flattened out by a blow without breaking,[2] by a process related to slip as we

[1] Martens-Heyn, "Materialenkunde," II, A, Springer, Berlin, 1912, p. 219; and Mügge, "Neues Jahrbuch für Mineralogie," 1898, Part I, p. 71.

[2] Beilby, *Journ. Inst. Metals*, 1911, No. 2, vol. 6, p. 13.

now understand it, and bismuth can be extruded as a wire,[1] quite as the crystalline coating of salt on the wire rod flows without losing continuity when the wire is drawn through the die, and as crystalline masses of marble can be moulded under strong pressure without breaking asunder,[2] the persistent pressure holding the little crystalline blocks together as they slide past each other under the great deformation which the collective mass undergoes.

This conception of the temporary mobile state helps us to understand these at first incomprehensible phenomena.

522. Continuity is Retained.—The course of slip which we have sketched in connection with Fig. 65, p. 337, suggests gaping. For instance it is a matter of common observation that the upper surface, *BD*, of a block thus punched becomes depressed saucerwise, so that the blocks immediately adjoining the path of the punch descend more than those farther away radially. But if block 12 descends less than 13, and yet all blocks retain their inclination unchanged, a cavity must form unless 7 crushes so as to compensate for this difference, its crushed particles following up the excess of descent of 13. But though this conception thus implies the formation of such cavities we are confident that none occur, because they could hardly be healed up by annealing, and yet annealing restores the initial properties of the metal almost exactly.

This conception of a temporary fluid state helps us to understand why cavities do not actually form. But the need of breaking up existing crystalline masses, in order to create this even temporarily fluid metal to fill up these crevices which tend to form, increases the resistance which the propagation of slip encounters at grain boundaries.

523. Why Increasing Crystallization Embrittles (§489).—It is sometimes denied that crystallization embrittles metal on the ground that, because metals are crystalline in any event, they cannot become crystalline. But this is to deny not the thoughts but the words, and is open to the full answer that they certainly can become more coarsely crystalline and probably more perfectly crystalline; that is that their crystalline organization can be made more perfect. The ideas which we have been considering show why crystallization in this sense really ought to embrittle, as common observation shows that it does.

Assuming, to fix our ideas, that slip occurs simultaneously along the whole of any one plane, and thus creates the weak mobile state along the whole of that plane simultaneously, coarseness of grain, and consequent great breadth and length of each such plane, may increase the proportion of the total potential path of rupture which is simultaneously weak. This sufficiently explains the brittleness which accompanies coarseness of grain. In the extreme case reported by Osmond, Fremont, and Cartaud,[3] a well-directed chisel blow along one of the cubic cleavages of an isolated crystal of

[1] Rosenhain, "Physical Metallurgy," Van Nostrand Co., New York, 1914, p. 262.
[2] Adams and Nicolson, *Phil. Trans. Royal Soc.*, 1901, vol. 195, A.
[3] *Rev. de Metallurgie, Memoires*, 1904, vol. 1, p. 45.

ferrite split it easily without deformation, whereas a blow at an angle of 45° to the cleavages cut it without breaking it. In the former case we may suppose that, because the blow created a mobile layer simultaneously throughout the whole cleavage plane, and thus extending completely along and across the crystal from end to end and from side to side, the two fractions which this layer separated simply fell apart, offering no resistance which could lead to their suffering deformation. I have found Hadfield's silicon steel extremely brittle when coarsely crystallized (§489).

Whether "vibration," which means only rapid variation or reversal of stress, can either coarsen or perfect the crystalline structure is a different question, which we can consider later.

524. Objections to Beilby's amorphous theory, to thus explaining plasticity as resulting from the passage through a mobile phase intermediate between the initial crystalline state and the final amorphous or undercooled liquid state, are first that the crystalline state is more stable than the amorphous so that the amorphous should always be tending to shift over into the crystalline state, and that a reversion from the crystalline to the amorphous state is contrary to the principles of thermo-dynamics. Second, that because the amorphous stage is bulkier than the crystalline, the effect of pressure as such should be away from the amorphous toward the crystalline state, on general principles.[1]

The objections do not seem well taken. The evidence that the surface of a crystalline mineral is made amorphous by polishing is extremely strong. If an outer surface can thus be amorphized by the friction of the polishing agent, no reason appears why internal surfaces along which slip occurs may not in like manner be amorphized by the friction of that slip. Opinions may differ as to how this reversion from a more to a less stable state occurs. Possibly it may be due to the tension, that is to the minus pressure, which arises in polishing an outer surface. Such decrystallizing tension on certain particles at certain stages of the internal slip may occur, and there may be no opportunity for recrystallization. That is to say, though the fingers press down on the surface which they are polishing, creating a vertical plus pressure, they may set up a horizontal drag, tension, or minus pressure, which is the true cause of the amorphization. Or, more broadly, the mechanical severing of the bond between the component parts of a crystal unit or space lattice may not be succeeded by an opportunity for a resumption of that or a like bond. It is rash to argue that, because the conditions of pressure on the mass as a whole ought not to cause it to amorphize or to revert to the liquid state, therefore they cannot cause such reversion along the internal slipping surfaces, because this reasoning assumes that the conditions throughout the mass are alike, which is untrue. The very anisotropy of a crystalline mass may well cause anisotropy of conditions under deformation, so that the conditions along the slipping planes may, in certain respects, be the oppo-

[1] Compare Tammann, *Zeits. anorganische und allgemeine Chemie*, 1915, vol. 92, p. 37; Hasselblatt, *idem*, 1915, vol. 93, p. 75; and Rosenhain, *Engineering*, 1913, vol. 96, p. 509.

site of those of the mass taken as a whole. The phase rule refers to conditions of equilibrium, and gives little support for predictions as to what can or cannot occur under conditions of inequilibrium, such as may well arise during this plastic deformation.

525. Alternative Explanations of Plasticity.—Heyn,[1] admitting that, as the plastic deformation deforms the individual grains, it creates on their surfaces of contact films which differ in their nature from anything present in undeformed metal, yet regards these films not as amorphous, not as differing from the rest of the metal or from undeformed metal in phase, but instead in surface energy. That is to say, the surface energy of small grains or of deformed grains is greater than that of large grains, with the result that small grains are always tending to unite and become large, and deformed ones to become equiaxed. Such variation in surface energy might explain the changes in physical properties which plastic deformation causes, but thus far it does not, so far as I can see, explain the phenomena of slip, that is to say the retention of continuity during the transition period, when each of the little particles *a*,*b*,*c*,*d*, etc., of Fig. 82 has parted company with its original vis-a-vis *a'*,*b'*,*c'*, or *d'*, and has not yet entered into relations with its new partner *c'*,*d'*,*e'*, or *f'*.

Tammann[2] finds that, during plastic deformation, the crystalline grains become subdivided into little polyhedra or thin lamellæ through the formation of gliding planes, which glide past each other, and holds that the formation of these gliding planes is the cause of the plasticity. Here, as in the preceding paragraph, I fail to find any true explanation, anything which makes the retention of continuity less miraculous, or connects it with familiar phenomena. This retention of continuity, it seems to me, is explained fairly well by Beilby, but by nobody else thus far.

Moreover, though most of the various changes which plastic deformation makes in the physical properties, the transparency of the surface film, the increase in strength, hardness, solution pressure, heat of solution, and volume, are just those which would naturally be expected to follow the creation of such amorphous metal as Beilby postulates, it is not clear that they ought to be caused by the agency imagined by either Tammann or Heyn.

526. Does Beilby's Theory Require that the Amorphous Metal along the Slip Planes shall cause Etching Traces?—If the hardness caused by deformation represents amorphousness, should the presence of that amorphous metal, forming the contact along every slipping plane, be detectable by etching? Beilby infers from his observations on the thickness of the amorphous layer formed by polishing a metallic surface, that the thickness of the supposed amorphous layer along the slip planes, when well-formed lamellæ slip over each other without plowing action, should be of the order of 100

[1] Martens-Heyn, "Materialenkunde," II, A, Springer, Berlin, 1912, §§264, 271 to 291, 293 to 313. Also *Proc. Internat. Assoc. Testing Materials*, VI, Congress, N. Y., 1912, 1st Section, II, 1, p. 43.

[2] *Op. cit.*, p. 46.

molecules, or say 30 $\mu\mu$, or far beyond the power of detection by means of the best lenses, though he believes that repeated slip causes layers thick enough to be seen.[1] If the layers, instead of being normal to the plane of the micro-section, were strongly inclined to it, they would indeed appear correspondingly thicker, provided that their full section persisted through the polishing and etching. Yet the outcrops of such inclined layers might well be eaten away by the etching and thus escape observation. The shadow cast by the furrow left by the removal of such very thin lamellæ might be undetectably narrow.

On the other hand deep etching might well be expected to cause a groove very much wider than the amorphous lamellæ themselves, as it does along the Neumann lamellæ, and as it widens out the grain boundaries. The failure of etching to develop traces of slip after moderate deformation is certainly a surprise. A still greater one is the weakening of the etching traces of slip by the very rest and gentle heating which simultaneously increase the hardening effect of that same slip (§§625, 626, and 634).

527. Why Repetitive Plastic Deformation Causes Anisotropy of the Elastic Limit.—After repetitive plastic deformation, for instance that of the endurance test in which slight permanent sets in tension and compression alternate with each other, the elastic limit is anisotropic.[2] If the final stress is tensile, then on releasing it and straining again tensilely the new elastic limit may exceed the initial greatly and may equal that last tensile stress, whereas if that new strain is compressive the elastic limit is not only less than the initial but may be zero. If we call the range between the elastic limit in compression and that in tension "the elastic range," the position of this range, as distinguished from its width, can be shifted very greatly by varying the range of those repetitive stresses, so as for instance to raise the tensile elastic limit very greatly, say to the initial tensile strength, lowering the compressive elastic limit correspondingly.

It is this anisotropy that Humfrey[3] explains ingeniously. Following out Beilby's amorphous theory, we must first assume that the elastic limit of the amorphous metal is extremely high, far beyond that of the crystalline. Conceive that, at an advanced stage of the endurance test, two small crystalline blocks E and F of Fig. 84 within any grain have, by the plastic deformation, become separated from their neighbors by a casing of amorphous metal *aaa*. Let later tensile stress in the direction of the arrows shift F to F' by sliding it along the slip plane rr', simultaneously generating fresh amorphous metal there, as shown at a' in Fig. 85. The state of affairs while this stress continues is shown by the broken lines in Fig. 85, E and F, which are rectangles when unstrained, bending under the existing tensile stress into rhombs,

[1] G. T. Beilby, private communication, Mar. 26, 1915.

[2] Unwin, "Testing of Materials of Construction," Longmans, London, 1910, p. 118. Bairstow, "Collected Researches," National Physical Laboratory, 1910, vol. 6, p. 180; and *Phil. Trans. Royal Soc.*, 1911, A, vol. 210.

[3] *Carnegie Memoirs, Iron and Steel Inst.*, "Influence of Intercrystalline Cohesion upon the Mechanical Properties of Metals." 1913, vol. 5, pp. 97–8. I give a very free rendering of Mr. Humfrey's hypothesis.

their amorphous casing a bending with them. Part of the tensile stress which has changed the shape of Fig. 84 into that of Fig. 85 has been relieved as regards the crystalline bocks E and F by their sliding past each other, but none of that in the amorphous sheath has been, with the result that the present residual tensile stress in a' is greater than in E and F. Let the external tensile stress now be released. Were E and F alone they would resume their natural rectangular shape; but when they have returned to that shape there is still residual stress in a', which now acting as compressive stress, will bend them back to the shape shown in unbroken lines in Fig. 85. At this moment E and F are under compressive and a' under tensile stress.

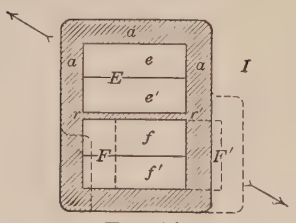

FIG. 84.

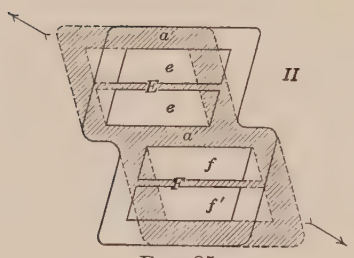

FIG. 85.

FIGS. 84 and 85.—Model to explain the monotropic effect (anisotropy) existing after repetitive stress.

If new tensile stress is now applied it will indeed tend to break E and F into their fragments ee' and ff' quite as E was first separated from F by slip along rr'. But the stress needed to start this slip is increased by the existing compressive stress which has brought E and F into their present rhomb shape. In order to slip e' past e and f' past f we shall have to exert, first enough tensile stress to bring them back to their normal rectangular shape, and then in addition the stress which should have been needed to effect that slip if they had been rectangular to start with, that is if that initial compressive stress had been lacking. Thus because the stress under which the preceding plastic deformation ended was tensile, the elastic component of that final stress, on the release of external stress, throws the crystalline parts into compression, and leaves the amorphous envelopes in tension. This compressive stress in the crystalline parts acts as a direct increase of the new tensile stress which will be needed to start a new slip, which starting is in itself the passing of the elastic limit. In short, the elastic component of the final tensile stress of the preceding phase raises the tensile elastic limit of the succeeding phase, which is what we started out to prove.

To change the presentation, given the initial state of compressive strain as distinguished from stress in the crystalline blocks, and the great excess of tensile elastic limit of the amorphous over the crystalline metal, though new tensile stress will cause new strain at the normal rate, the first part of that strain of the crystalline metal will be consumed in removing its existing compressive stress. The strain up to this point increases the total strain at which the elastic limit of the crystalline blocks will be reached.

Like reasoning shows that that elastic component decreases the compressive stress needed to cause f' to slip compressively past f and e' past e, explaining why a final tension lowers the elastic limit in compression. Thus the anisotropy is explained.

Should the addition of the new tensile stress to the residual break the amorphous casing, a', though this would lessen it would not remove the spring action, because of the integral union between the casing and its crystalline contents.

In Humfrey's original description the amorphous metal is supposed to form envelopes, not about the little crystalline blocks within each grain, but about the grains themselves, following his hypothesis (§701) that there is a progressive departure from crystallinity as we pass from the interior of the grains to their boundaries. In view of the reasons given in §§702 *et seq.* for questioning this hypothesis, I have sought to get the benefit of his conception of greater elasticity of the amorphous than of the crystalline metal by adapting it to what seems to me the more reasonable belief, that each fraction of slip subdivides the crystalline blocks, encasing their fragments with amorphous metal, as in Figs. 84 and 85.

528. Why Anisotropy after Repetitive Plastic Deformation is Replaced by Isotropy after a Single Deformation.—The effect on the elastic limit of a single plastic deformation as distinguished from repetitive deformations is isotropic, as great in compression as in tension. Immediately after the deformation the elastic limit may be very low or even zero, not only in compression but in tension. During rest it seems to rise equally in tension and compression.[1]

The contrast between the isotropy in this case and the anisotropy in the preceding may throw light on the life history of the amorphous metal. The suggestion is that the difference represents a change which occurs in the properties of this metal during the long period of the repetitive deformation.

The isotropy which exists after a single plastic deformation may be explained thus. In the brief time here involved there are some parts of the amorphous metal, those for instance last generated, which reach only very moderate strength. If after the deformation the metal is re-strained at once, permanent set occurs in these weak parts while the new stress is still very gentle, and this is true whether this new stress is tensile or compressive. Hence the lowering of the elastic limit. Because of the weakness of this lately formed amorphous metal, it retains no important residual stresses such as are suggested by Figs. 84 and 85. Hence the isotropy.

During rest or gentle heating the amorphous metal sets and gains progressively in strength. Hence the gain of the mass as a whole in hardness and elastic limit. But because there was no important residual stress at the end of the first plastic deformation, there is no important cause of anisotropy. Hence the isotropy even after rest. The essential difference imagined between these two cases is that between the existence of great residual stress after the repetitive overstrain, because of the high elastic limit which the amorphous metal has reached in the long period since its generation began,

[1] Thomas Gray, *Trans. Amer. Inst. Min. Eng.*, 1897, vol. 18, pp. 707–9. Unwin, "Testing of Materials of Construction," Longmans, London, 1910, p. 118. Galy-Aché, *Rev. Metallurgie, Mem.*, 1913, vol. 10, p. 587. The Author, *Proc. Amer. Soc. Testing Materials*, 1914, vol. 14, p. 7.

and the absence of such residual stress after a single overstrain because of the low elastic limit which the amorphous metal reaches in the brief time since its generation.

Further light on the properties of amorphous metal may be thrown by the contrast between the great lowering of the elastic limit immediately after a single plastic deformation and the simultaneous rise of the yield point. The permanent set between the elastic limit and the yield point we may conceive to occur in the amorphous metal itself, whereas at the yield point there begins a new period of great breaking up of the crystalline metal, and consequent great stretch.

529. Present Position of the Amorphous Theory.—This brilliant theory is extremely useful in explaining a great mass of extraordinary and hitherto unexplained phenomena. The doubts which remain today, so soon after its enunciation, as to its competence to explain the intricate phenomena discussed in the last three sections are indeed such as might be expected. They may lead us to class it rather among the precious working theories than among those firmly established.

Closer study of the properties of the amorphous metal is certainly needed, with a view to clearing up these points, and especially such paradoxes as the weakening of the etching traces of slip by heating and rest, though these intensify the hardening effect of that same slip; the contrast between the isotropy after a single plastic deformation and the anisotropy after repeated ones; and the contrast between the apparent rapid strengthening of the amorphous metal indicated by the rapid rise of the stress-strain diagram and of the yield point on one hand, and the weakness of at least part of that metal indicated by the simultaneous lowering of the elastic limit on the other hand. The speculations which I have offered in these sections are, of course, intended only to incite others to those direct observations by which alone can the truth be reached.

CHAPTER 21

PLASTIC DEFORMATION IN STEEL

530. Introduction.—In the preceding chapters we have considered the phenomena of plastic deformation by slip in pure ferrite, in which all the conditions are the most simple because, though it is composed of distinct crystalline grains, all these grains are of one and the same substance. Let us now consider plastic deformation in the conglomerates of ferrite with pearlite of which our industrial steels consist.

531. The Plastic Deformation of all Kinds of Iron is Primarily in their Ferrite.—The cementite present, whether pearlitic or pro-eutectoid, may break or bend, but it probably does not yield plastically to any large degree.

Though the path of rupture itself shows but little preference for pro-eutectoid ferrite over pearlite, as we shall see in §715, yet the early stages of deformation probably occur chiefly in the pro-eutectoid ferrite. Let us look at some evidence supporting this supposed law.

When the pro-eutectoid ferrite of hypo-eutectoid steel is concentrated into large masses, as in the Widmanstättian figuring of Fig. B of Plate 28, prominent slip bands may be developed in it, such as are shown in Fig. D of that plate.

Even when the pro-eutectoid ferrite is assembled in much smaller masses, it yields in such a way as to accommodate itself to the deformation, and thus enables the mass as a whole to retain continuity while changing shape. For instance a ferrite network which has been developed in a medium carbon steel by cooling at an intermediate rate from a high temperature well above *A*3, emerges above the previously polished surface when the steel is compressed, and may then be recognized readily by its brightness under oblique illumination, as in Fig. E of Plate 27. Rotation through 180° brings back the same bright lines, thus proving that they are not steps but either ridges or furrows, and in this case evidently ridges.

This interpretation is supported by the appearance of this same specimen on greater magnification, shown in Fig. F of Plate 27. This, too, shows an unetched surface polished before the deformation, but seen here under direct instead of under oblique illumination. Here the uplift of the ferrite network along *ab* can be seen clearly.

The behavior of a similar but thinner ferrite network *DE*, *FE*, *EG*, in a steel of 0.40 per cent. of carbon shown in Fig. B of Plate 27,[1] is interpreted

[1] This same specimen after etching with picric acid is shown in Fig. E of Row 6, Plate III, of my paper "Life History of Network and Ferrite Grains in Carbon Steel," *Proc. Amer. Soc. Testing Materials*, 1911, vol. 11, following p. 386.

less easily. It shows here as a narrow black band separating the three sorbitic pearlite kernels. With Figs. E and F of Plate 27 before us, we interpret this band as an uplift of ferrite so roughened and rounded that it reflects but little light vertically into the microscope or camera tube.

Though the slip bands cannot actually be seen in such narrow masses of ferrite, yet the mechanism by which the ferrite is enabled thus to move may reasonably be assumed to be slip like that by which it moves when in larger masses.

532. The reason why I infer that the deformation here is primarily in the ferrite is that this is the only plastic substance present. The other constituents of these alloys, such as cementite, slag, sulphides, and graphite, are so extremely brittle that they probably lend themselves only very reluctantly and sparingly to plastic deformation. Thus Fig. B of Plate 26 shows how the large cementite masses in a very high-carbon steel break under distortion into large vertebra-like islands, bathed in a sea of pearlite.

The illustrations given in the last section showed the plastic deformation of pro-eutectoid ferrite. But the very considerable capacity for plastic deformation of eutectoid steel of 0.90 per cent. of carbon, consisting wholly of pearlite, must be due primarily to its pearlitic ferrite. The pearlitic cementite must be as brittle as the pro-eutectoid. The fact that deformation which might be expected to break this brittle pearlitic cementite does not necessarily break the mass as a whole must be referred to the great excess of pearlitic ferrite in which this cementite is bathed, each part of cementite being accompanied by some 6½ parts of ferrite.

Under strong compression the brittle cementite and slag may perhaps retain continuity even when crushed, quite as the brittle antimony, salt, and marble can (§521). But in general even under such compression the ferrite present, both pearlitic and pro-eutectoid, probably plays an essential part in holding the crushed particles together so that they may retain or regain continuity.

The reasons which will be given in Chapter 27, tending to show that the movement in ferrite may be in part fluid, apply also to both the pro-eutectoid and the pearlitic ferrite of steel.

This slip of the ferrite is likely to be much greater at its contact with the masses of cementite, whether pearlitic, pro-eutectoid, or other, than elsewhere. These brittle cementite masses are probably forced through the ferrite like a punch, which certainly causes much greater deformation along its contact with the metal than elsewhere, as shown in Fig. A of Plate 18. That the slip in the pearlitic ferrite is thus chiefly along its contact with the pearlitic cementite tallies with the behavior of the lead-tin eutectic. In this eutectic the deformation occurs chiefly by means of slip in the light colored component which corresponds to the pearlitic ferrite, along its contact with the other component, which corresponds to the pearlitic cementite.

That the slip in a constituent of a eutectic or eutectoid is truly crystalline is shown in the case of the bismuth-lead eutectic by the parallelism of the slip

bands in separate patches of the slipping or light constituent, which no doubt were connected outside of the plane of the microsection.[1]

The reason why I infer that plastic deformation in its early stages is chiefly in the pro-eutectoid rather than in the pearlitic ferrite is that the incomparably greater grain size of the former probably makes it very much weaker than the pearlitic ferrite (§§479 *et seq.*), as indeed we may infer from the much greater elastic strength of pearlite, as we know it in eutectoid steel, than of nearly pure ferrite as we know it in ultra low-carbon steel.

533. That the Slip is in the Ferrite Helps to Explain the Influence of Carbon on the Tensile Properties.—*The elongation*, that is, the power of enduring plastic deformation without rupture in the tensile test, decreases as the content of free or pro-eutectoid ferrite decreases from 100 per cent. in pure iron to zero in 0.90 per cent. carbon or eutectoid steel, as shown in Fig. 10, §69. With further increases of carbon content and corresponding decrease in the proportion of pearlite and hence of pearlitic ferrite, the little remainder of ductility further decreases progressively.

This tallies exactly with the belief that, in hypo-eutectoid steels, the ductility is given chiefly by slip in their pro-eutectoid ferrite, and secondarily by that in their pearlitic ferrite; that in hyper-eutectoid steel and white cast iron it is given almost if not quite solely by slip in their pearlitic ferrite; and that in gray cast iron it is given by the corresponding ferrite in the metallic matrix, through which their graphite is scattered as explained in §117.

534. The Tensile Strength and Elastic Limit. The Obstruction Theory.—The usual explanation of the increase of the tensile strength and elastic limit with the carbon content is, briefly, that the stronger pearlite stiffens the mass by impeding the flow of the weaker ferrite; and, going a step further, that the reason why pearlite, though made up of about 6 parts of ferrite to 1 of cementite, is yet so much stronger than pure ferrite, is that this pearlitic cementite thus stiffens the pearlitic ferrite. The pearlite in one case and the pearlitic cementite in the other, is said to "hold up" the ferrite. After discussing the "obstruction principle" I call attention to the "ferrite-refinement" principle, that the effects of increasing carbon content may be due in considerable part to the accompanying changes in the grain size of the ferrite.

The case often used to support the obstruction theory, the strengthening of the walls of a clay crucible by means of a coke skeleton, may be objected to as unfair. Here the shape of the tongs is such that the crucible walls are nearly in the condition of a uniformly loaded arch, and the volume of coke is so great that, under this pressure, coke probably presses directly against coke. The fact that coke, which is perfectly rigid at the high temperature of the crucible while the clay itself is very soft, gives com-

[1] Ewing and Rosenhain, *Phil. Trans. Royal Soc.*, 1900, A, vol. 193, pp. 371–2; and *The Metallographist*, 1900, vol. 3, pp. 127 to 129. Also Rosenhain, "Physical Metallurgy," Van Nostrand Co., N. Y., 1914, pp. 260–1.

pressive strength, hardly leads us to expect with confidence that scattered coke particles should give tensile strength to a mass of plastic clay. My preliminary experiments do not indicate that the strength of wax, even under compression, is increased greatly by mixing either small or large proportions of sand with it, perhaps because the contact was not intimate enough.[1]

But apart from this simile we can understand that the presence of any hard unyielding body within the ferrite may stiffen it to an important degree by opposing the slip along its slip planes, as I will now explain.

Assume that the existing stress tends to cause slip in the direction of the arrow in Fig. 86 along the slip plane *CD* of a given ferrite grain. If this grain abutted on its side *DE* against another ferrite grain, the disregistry between these two grains would indeed oppose the slip along *CD* as we have already seen; but the resistance which the still harder and therefore stronger martensite offers to slip along *CD* is proportionally greater, somewhat as the resistance of a snow shoe is greater than that of snow. The tendency to slip along *CD* concentrates pressure upon the point *D*. If *D* rested against another ferrite grain, only the strength of this second grain would resist that slip. Resting against martensite, the greater strength of the martensite resists that slip correspondingly more strongly.

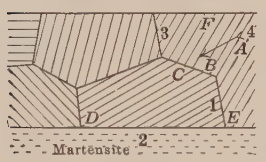

Fig. 86.—Foreign bodies oppose slip.

Thus if flow occurred only by the sliding of one grain as a whole over another grain, then the resistance of the flow of the ferrite grain 1 along the martensite grain 2 should be much like that of one ferrite grain over another, and the strengthening effect of the martensite would be hard to understand. But because movement occurs through the slip of one ferrite stratum over another within the various ferrite grains, we readily understand how the greater resistance of martensite than of ferrite to this intragranular slip should strengthen the mass.

The hardness of the martensite prevents slip from occurring in the ferrite grain along *CD*, and thereby in turn opposes the tendency of the slip to occur along the plane *FC* in grain 3, and so starts a resistance which in effect propagates back from grain to grain, as a down-stream dam raises the water level far up-stream.

In addition to this the presence of the martensite certainly offers frictional resistance to the displacement of deformation, as the presence of any solid body does to the flow of a liquid or of wax. It is true that this frictional resistance is likely to decrease with the rapidity of movement, and thus to be relatively slight in the slowly-moving tensile test.

535. Is the Obstruction Theory Competent Quantitatively to Explain the Increase of Strength Caused by Increasing Carbon Content?—In order

[1] Abraham finds that the compressive strength of bitumen is increased by the intermixture of mineral matter, but his remarks do not point to an increase of tensile strength of an order of magnitude approaching that which occurs in carbon steel with increasing carbon content. (Private communication, and *Trans. Amer. Soc. Testing Materials*, 1911 vol. 11, p. 698.) Kirschbraun's observations (*loc. cit.*) do not conflict with this inference.

to answer this we may pass at once to an extreme case, that in which steel of 0.21 per cent. of carbon, when quenched in water from within the transformation range, had a tensile strength of 142,000 lb. per square inch,[1] nearly thrice that of pure ferrite, though it contained some 60 per cent. of ferrite. The microstructure of this specimen is shown in Fig. 87.

Under tensile stress in the direction of the arrows in this figure, rupture could avoid all the black martensite islands and pass wholly through the ferrite from *E* to *F*, and this would lengthen the path of rupture by only 4 per cent. Can so slight a lengthening of the path of rupture be the true cause of this tripling of the strength of the mass?

Of course some allowance must be made for the fact that what here look like scattered martensite islands are in fact connected outside the plane of the section, so that the real structure is more like that of a wax ingot cast in a mould into which some pine trees of martensite had been stuffed. May not the strength of the martensite skeleton masses contribute greatly to the strength of the whole?

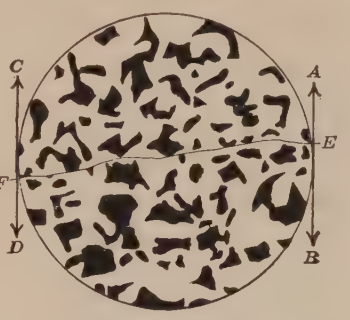

Fig. 87.—The probable path of rupture in low carbon steel. Howe and Sauveur.[1]

In order to answer this let us consider what occurs during the tensile test. The ferrite has great capacity to flow, the martensite has very little. When the stress, as spread out over the conglomerate, reaches the yield point of the ferrite, this ought to yield, and thus to throw an ever-increasing proportion of stress on the unyielding martensite skeleton, which therefore ought soon to break, considering how very crooked this skeleton is, how unaxially it meets the stress, what local thinnings its section has, and what sharp re-entering angles.

Because the martensite is incapable of taking any considerable permanent set, this skeleton must break up into disconnected fragments before the stretch has gone far, and hence it can no longer act as a skeleton to strengthen by its own continuity the composite mass as a whole, any more than reinforcing steel bars could thus strengthen a mass of concrete skeletonwise if they were reduced to disconnected boiler punchings. And yet after the elastic limit has been passed in the tensile test, and this initial martensite skeleton has thus been broken up into disconnected vertebræ, the composite mass continues growing stronger, as is shown by the progressive rise of the stress strain diagram up to the maximum load, in spite of the progressive loss of sectional area of the test piece.

Frankly, the explanation, that this specimen is thrice as strong as pure ferrite because the martensite obstructs the flow, is not so fully satisfying

[1] Howe and Sauveur, "Further Notes on the Hardening of Steel," *Journ. Iron and Steel Inst.*, 1896, No. I, p. 177, and Fig. 6, plate VIII. We pointed out in this paper that the strength of this specimen was inexplicable by any of the theories then current.

as to prevent our turning gladly to any contributory explanation, such as that of the "ferrite-refinement" theory.

536. **The ferrite-refinement theory of the increase of tensile and elastic strength with the carbon content** helps to explain the great strength of this specimen. First, it was quenched soon after cooling into the transformation range, and therefore before its pro-eutectoid ferrite, generated on thus cooling into this range, had had a chance to grow into coarse grains. Second, the ferrite is broken up by the martensite skeleton in such a way that its growth into large grains is impeded. Both these things make the ferrite fine grained, so that it has a great number of grain boundaries. Third, the inter-stratal movements, due to the variations from stratum to stratum in the rapidity of cooling and hence of contraction during the quenching, might well induce local plastic deformation such as, in lines 16 to 21 of Table 20B, §489, led to a marked excess of strength of the finer over the coarser ferrite.

In view of the sluggishness of the material at this temperature, not only may the absolute grain size be extremely minute, but the elaboration of the crystalline structure, the development of the weakness of the slip planes, may be very incomplete. At least here are agencies which may contribute to an important degree to the extraordinary strength of such specimens. It is possible that the influence of reducing the ferrite grain size from that common in low-carbon steel to the submicroscopic dimensions of the pearlitic and sorbitic ferrite may be out of proportion to that of the variations in grain size with which Tables 18, 19, and 20A and 20B have to do.

Other contributory influences are the foreign elements dissolved in the ferrite itself, and the shape and size of the cementite masses. Indeed these last are probably of the first importance in the case of white cast iron, by determining whether the cementite shall break under given deformation and also the size and harmfulness of the cracks which that breaking creates.

On the other hand, the increase of the tensile strength with increasing carbon content in hypo-eutectoid steel, as shown in Fig. 10, is not readily explained thus, because here the differences in the size of the grains of the pro-eutectoid ferrite, caused by the increasing quantity of pearlite, are of the order of those which underly Tables 18 to 20B, and because they there have a relatively slight influence on the strength. Here, therefore, we refer the increase of strength primarily to the obstruction principle.

537. **In Hyper-eutectoid Steel and Cast Iron.**—When the carbon content rises beyond the eutectic ratio of 0.90 per cent., the strength indeed usually decreases. This is referable to the introduction of increasing quantities of pro-eutectoid cementite, itself incapable of yielding, and hence, if assembled in large masses, tending to break up and to start cracks too large to heal by the inflow of the yielding ferrite, and thus to precipitate rupture. But if this cementite is prevented from coalescing into large masses so shaped as to be likely to break, and is scattered as a fine emulsion through the ferrite, which in turn is thus broken into an emulsion and hence has ultra-micro-

scopic grains, the strength, both elastic and maximum, may increase till the carbon rises far beyond the eutectoid ratio.[1]

The undoubted tendency of the strength to decrease as the carbon content rises above the eutectoid ratio is to be referred to the tendency of the pro-eutectoid cementite to assemble into masses so shaped as to be likely to break, and to start cracks too large to be healed. When the carbon content rises above 1.70 per cent., the eutectic cementite formed during soldification enters, almost necessarily in masses so large as to be harmful in this very way.

This we conceive the more readily after inspecting the microstructure of an excellent 2.04 per cent. carbon steel, Fig. B of Plate 26. These great white islands are cementite which has broken up into vertebral fragments. We can readily understand the difficulty with which the pearlite surrounding these islands can be forced into the minute spaces between them, and hence the slight ductility of this steel when below the transformation range.

538. Other Cases in which the Grain Size of the Ferrite may be Important.—It remains to be seen whether the merit of fineness of grain may help to explain many other phenomena otherwise obscure, such as the fact that eutectoid steel, which consists of pearlite alone, may be nearly thrice as strong as ultra low-carbon steel which consists of ferrite almost alone, though six-sevenths of that pearlite is made up of ferrite differing from the ferrite of the low-carbon steel primarily in being far finer grained; the greater strength of the sorbitic than of the pearlitic structure whether brought about by heat treatment, by the presence of say 1 or 1.25 per cent. of manganese, or by means of vanadium; and, on the other hand, the harm done by divorcing annealing and other coarsening processes.

539. The effects of divorcing annealing suggest that they may be due in part to the ferrite-refinement principle. Thus the degradation which divorcing annealing causes is roughly proportional to the pearlite content, and hence to the quantity of pearlitic ferrite which this annealing frees from its interweaving in submicroscopic grains with the pearlitic cementite, and thereby allows to coalesce into incomparably large grains along with the pro-eutectoid ferrite. This is shown in Fig. 88 and Table 21.

[1] Thus Roberts-Austen found that both the elastic and the maximum tensile strength increased with the carbon content at least up to 1.30 per cent., when initial differences of finishing temperature, etc., were lessened while the structure was kept fine by heating to 620°. *Sixth Rept. Alloys Research Comm.*, 1904, excerpt *Proc. Inst. Mechan. Engin.*, Fig. 14, Plate 7. My own data, which will be given in another volume, prove this abundantly.

Note to Table 21.—Previous Treatment.—All the specimens of Nos. 1, 2, 3, 4, 6, and 7 were preheated to 900°; the "initial" specimens were then air cooled to room temperature and the "final" ones furnace cooled to between 680° and 700° and held there 24 or 96 hours. No. 5 was held at between 680° and 700° for 1½ hours before the initial properties were determined. For Nos. 8 and 9 the "final" specimens were held between 650° and 700° for 19 days without any previous treatment, and are compared with the "initial" of Nos. 2 and 4, which were air cooled from 900°.

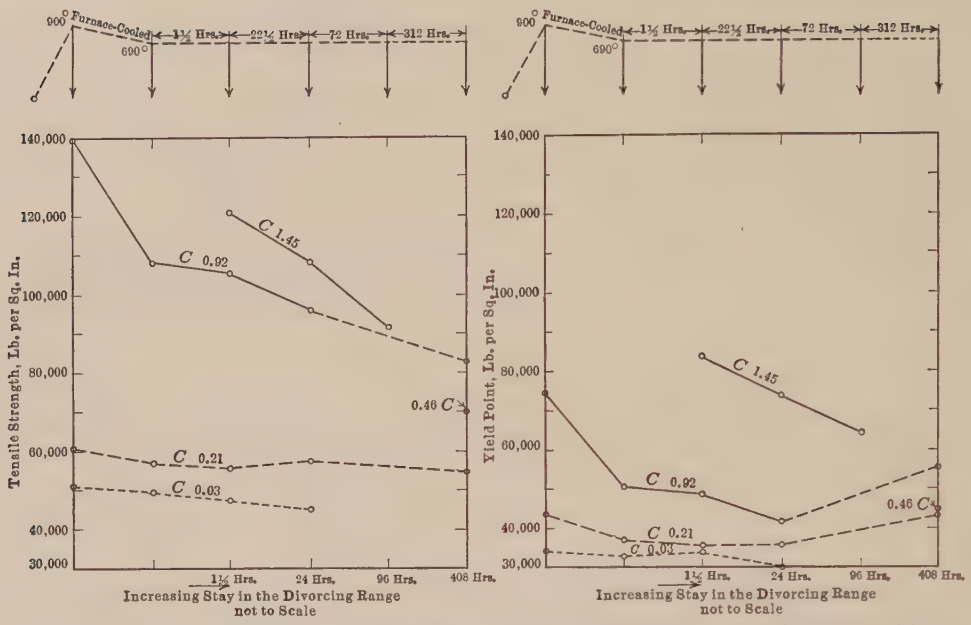

FIG. 88.—The degradation caused by the divorcing annealing of pearlite increases with the carbon content.[1]

NOTE.—The Brinell diagram at the top does not apply to the 1.45 carbon curve, or to the 408-hr. points.

TABLE 21.—THE WEAKENING CAUSED BY THE DIVORCING ANNEALING OF PEARLITE INCREASES WITH THE PEARLITE CONTENT

Number	Carbon content, per cent.	Steel, No.	Properties before and after divorcing annealing										Loss (−) or gain (+) of properties by divorcing annealing in percentage of the initial				
			Tensile strength, lb. per sq. in.		Yield point, lb. per sq. in.		Elongation, per cent. in 4 in.		Contraction of area, per cent.		Impact resistance (Fremont test)		Tensile strength	Yield point	Elongation, per cent. in 4 in.	Contraction of area, per cent.	Impact resistance (Fremont test)
			Initial	Final	Initial	Final	Initial	Final	Initial	Final	Initial	Final					
			24-hour period										24-hour period				
1	0.03	2	51,040	46,550	34,690	30,010	25.9	27.12	75.5	75.75	2.4	4.5	−10.8	−11.9	+ 4.8	+ 0.3	+91
2	0.21	4	60,800	57,580	44,300	38,150	20.5	20.2	63.16	59.25	26.0	2.4	− 5.3	−13.9	− 1.5	− 6.2	−91
3	0.59	11	.....	.....	.....	.....	.....	.....	.....	.....	4.5	3.5	.....	.....	.....	.....	−22
4	0.92	14	139,425	96,300	75,420	41,870	7.25	10.5	16.39	19.2	2.6	0.6	−30.9	−44.5	+44.8	+17.8	−74
5	1.45	24	121,000	108,500	84,500	75,000	4.4	5.7	7.2	10.6	1.6	2.2	− 9.9	−10.7	+30.0	+40.0	+37
			96-hour period										96-hour period				
6	0.59	11	.....	.....	.....	.....	.....	.....	.....	.....	4.5	2.5	.....	.....	.....	.....	−44
7	1.45	24	.....	.....	.....	.....	.....	.....	.....	.....	1.6	2.5	.....	.....	.....	.....	+56
			19-day period										19-day period				
8	0.21	4	60,800	54,600	44,300	43,200	20.5	19.85	63.16	66.1	.....	.....	−12.0	− 2.5	− 3.1	+ 4.7	.....
9	0.92	14	139,425	83,080	75,420	55,250	7.25	16.25	16.39	39.75	.....	.....	−42.6	−26.7	+124.0	+142.6	.....

[1] Deduced from the data of Howe and Levy, "Notes on Divorcing Annealing," *Proc. Cleveland Inst. Engineers*, 1913–14, p. 218 (July) with the exception of the 408-hour data.

In interpreting these results we must allow for the much slower divorcing annealing of higher than of lower carbon steel.

Here ultra low-carbon steel which is substantially free from pearlite undergoes only the slight degree of degradation readily referred to the growth of the grain size of the pro-eutectoid ferrite of which nearly the whole of this mass consists. The 0.92 per cent. carbon steel, on the other hand, undergoes in divorcing annealing the much greater degradation corresponding to the change of its ferrite from its initial pearlitic state, in which its grains are submicroscopic, to a free state in which the growth of its grains is relatively little impeded by the coagulated masses of cementite.

We must not overlook the damage which divorcing does by assembling the divorced cementite into harmfully large brittle atolls, such as are shown in Fig. A of Plate 26, which shows the structure of a divorced steel of 0.21 per cent. of carbon in the immediate neighborhood of the tensile fracture. The breaking up of these cementite islands might well cut gashes hard to heal, and thus start rupture.[1]

540. The tempering phenomena make a like suggestion. Thus Mathews and Stagg find that the loss of hardness, elastic limit, and tensile strength caused by tempering at 427° is very far from complete at the end of 20 minutes, though the chemical transformation from austenite into ferrite and cementite should have become nearly if not quite complete, as we infer from Barus and Strouhal's[2] finding that even at 330°, at which the change is much slower than at 427°, the transformation, as indicated by the change of thermoelectric power, reached almost the maximum corresponding to that temperature in ½ hour. That the transformation should have become complete in the experiments of Mathews and Stagg is further indicated by Maurer's data showing that the density, perhaps the most trustworthy measure of the transformation, reaches very nearly if not quite the maximum on tempering for 1 hour at some temperature between 380° and 450°.

The rapidity of the transformation is indicated further by Benedicks'[3] observation that retarding the quenching of steel of 1 per cent. of carbon from 725° in such a way that it occupied 9.5 seconds in cooling from 700°

[1] It is true that so eminent an observer as Stead seems to doubt whether the divorced cementite actually has this effect (*Fifth Report Alloys Research Comm.*, excerpt *Proc. Inst. Mechan. Engin.*, 1899, p. 73). Here he cites a "highly brittle annealed piece of soft Swedish steel containing 0.04 per cent. of carbon, showing the free cementite in small white patches. On breaking this steel, fractures passed through the ferrite and not through the brittle cementite." To this one may reply, first, that rupture which has been precipitated by the sharpness of any one of some hundreds of such pieces of cementite within the specimen and not visible on its surface would in only a very small proportion of cases happen to pass also across any one of the few cementite islands upon the surface or disclosed in a given miscrosection, the total proportion of cementite being so extremely small, only 0.6 per cent.; and second, that the extreme brittleness of this specimen might well determine rupture along a few broad cleavage planes of special weakness, whence it could not depart for the purpose of cutting any given superficial cementite island.

[2] "On the Physical Characteristics of the Iron-Carburets," *Bulletin* 14, *U. S. Geological Survey*, 1885, p. 55.

[3] "Experimental Researches on the Cooling Power of Liquids," *Journ. Iron and Steel Inst.*, 1908, No. II, vol. 77, pp. 218 to 221.

to 100° gave time for the completion of the transformation, so that the density and electric resistance then were identical with those of the annealed state. Indeed at 400° the transformation, instead of going on only to a certain degree as at considerably lower temperatures, seems to complete itself.

Table 22.—Progressive Softening of Quenched Steel at 430°, Mathews and Stagg

Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation		Reduction	Brinell	Oil quenched	Tempered
		Per cent.	In.				
260,137	228,750	2.5	2	.....	425	843°C.	427°C. for 8 min.
214,562	201,125	11.6	2	45.4	390	843°C.	427°C. for 20 min.
183,187	175,000	12.0	2	49.35	340	843°C.	427°C. for 40 min.

Note to Table 22.—One-half-inch round bars of steel of carbon 0.50, silicon 0.50, manganese 0.75, chromium 0.75 were quenched in oil from 843°C. (1,550°F.), and then drawn in a salt bath at 427°C. (800°F.) for the time shown. "Factors in Hardening Tool Steel," *Trans. Amer. Soc. Mech. Eng.*, Dec., 1914, vol. 35, p. 20 of preprint, and private communication, Dec. 8, 1914.

Inferring that prolonging the stay at 427° from 20 to 40 minutes did not induce any important degree of transformation in the experiments of Table 22, because this should have been nearly complete before this prolongation, I infer further that the great loss of hardness, 50 Brinell numbers, in this prolongation is probably due to the coalescing of the emulsified ferrite and cementite into larger though still microscopically irresoluble particles. A like inference is to be drawn from Heyn's showing that the loss of hardness of hardened steel is only 70 per cent. on tempering at 400°, and 87.5 and 97.5 per cent. on tempering at 500° and 600°,[1] and from Maurer's[2] finding that even in his specimens in which the density change induced by tempering reached a maximum at 450°, indicating that the transformation was complete, further heating to 500° and 600° caused a further softening, which might be due to this grain growth. See Fig. 123, p. 599.

The evidence of Mathews and Stagg is weakened somewhat by the presence of 0.75 per cent. of chromium, the disturbing effect of which we cannot estimate confidently.

These inferences are based on the belief that the changes in grain size in this submicroscopic range may have much greater effect than changes in the relatively coarse range experimented on in connection with Tables 18 to 20.

541. Intrapearlitic Deformation.—The cementite lamellæ, perfectly brittle as they are, can yet undergo great deformation without showing that they are broken, provided that the mass is held firmly together by pressure on all

[1] *Mitteilungen a. d. kgl. Materialprüfungsant, Gross Lichterfelde West*, 1906, No. 1; through *Stahl u. Eisen*, 1906, vol. 26, p. 780.

[2] *Rev. de Metallurgie, Mem.*, 1908, vol. 5, p. 718; *Metallurgie*, 1909, vol. 6, p. 33. See also Boynton, *Journ. Iron and Steel Inst.*, 1906, No. II, vol. 70, p. 287; and "The Shore Scleroscope," The Shore Instrument and Manufacturing Co., New York, Aug., 1912, p. 39.

sides, as for instance in punching. Note how they are crumpled into a W form in the right-hand part of Fig. F of Plate 2, and bent into a V or U at the lower left-hand end of this pearlite island, and at a point ½ in. E. from *b* of Fig. F of Plate 26. This flexibility should hardly surprise us when we remember that these thin lamellæ are only about 0.000,03 in. thick, and that even macroscopic hairs of glass and of slag wool can be bent greatly. This deformation is probably wholly elastic, quite as that of bent glass hairs is.

Under other conditions the cementite breaks up into fragments which shift their position without necessarily changing their direction materially, somewhat as sketched in Fig. 89. Here, starting with the disposition shown in 1, stress in the directions of either *A* and *B* or *C* and *D* may shorten and widen the ferrite lamellæ, and thereby move the cementite fragments into more open order *en echelon*. This arrangement can be traced in the left-hand part of the pearlite mass of Fig. F, Plate 2. Or stress in the direction of either E and F or G and H, lengthening and narrowing the ferrite lamellæ, may thereby move the lines of cementite fragments nearer together, though moving the members of each line farther from each other,[1] as at 3 of Fig. 89. This effect can be traced in the right-hand part of the pear-shaped pearlite mass of Fig. F of Plate 26 opposite *a* and lower down. At the base of the pear the cementite lamellæ are moved apart somewhat as in 2 of Fig. 89, while toward its neck they are arranged more as in 3 of that diagram.
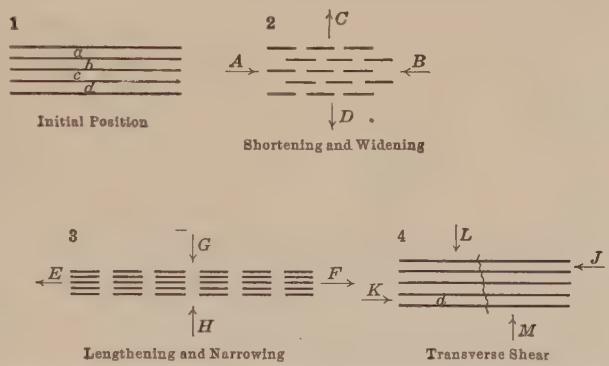
FIG. 89.—Conjectured behavior of cementite laminæ alternating with ferrite under deformation.

Rupture, according to Rosenhain,[2] passes oftener across than along the pearlite stratification, as is natural. Shear caused by the stresses *J* and *K*, in 4 of Fig. 89, can occur along any one ferrite lamella, for instance *d*, without breaking the ductile ferrite, whereas shear across the stratification, as at *LM*, must soon break the inextensible cementite. The incipient rupture thus started may develop. Thus a deformation which cuts across the stratification is more likely to lead to rupture than one which runs parallel to it, in accordance with Rosenhain's observation.

542. Deformation Lines on the Previously Polished Surfaces of Pearlite.—If we cannot detect slip bands in pro-eutectoid ferrite when it is in small masses, we cannot expect to see them in the pearlitic ferrite, much less in

[1] See Howe and Levy, "Notes on the Plastic Deformation of Steel During Overstrain," *Trans. Amer. Inst. Min. Eng.*, 1914, vol. 50, p. 542.

[2] "Deformation and Fracture in Iron and Steel," *Journ. Iron and Steel Inst.*, 1906, No. II, vol. 70, p. 220.

sorbite. Nevertheless, deformation in both probably occurs through this slip in their ferrite. But though the individual slip bands in pearlitic ferrite cannot be detected, the uplifting which deformation causes in this ferrite as a whole can be seen very clearly on the unetched surface of a specimen polished before the deformation, as in Figs. A and F of Plate 27. Here the ferrite lines stand up prominently, creating a lamellar figuring closely like that caused by etching, though less sharply defined.[1]

543. Deformation Lines on the Previously Polished Surface of Sorbite.— Here deformation creates lines which look much like slip bands. Note the fine hairy groups of lines in Fig. B of Plate 27 below *A* and *C* and at the left of *B*. At the right of *C* in Fig. C of Plate 27 note the vermicelli lines running N-55°-E and resembling those of Fig. D of Plate 28. On the other hand, the great scimetars such as *aa*, *b*, and *ccc* in Fig. D of Plate 27 are more like the thorns of Osmond (§583).

The meaning of these lines is yet to be studied. They can hardly represent slip along cleavage planes in this very fine emulsion of ferrite and cementite, because we can hardly suppose that the submicroscopic particles of these constituents really have the common orientation over these relatively large distances which continuous slip planes would imply. Hence the inference that they are probably foldings representing fluid motion. This, and their strong resemblance to many of the lines in pure ferrite which at first seem to be slip bands, tend to suggest that the true slip bands, even though representing slip along cleavages, may yet have something of the fluid in their movement, somewhat as the crystalline nature of liquid crystals is no bar to their incipient spheroidizing.

The upheaval along what has been an austenite grain boundary can be followed on a larger scale in Fig. F of Plate 27, from *a* to *b*, on a surface of a sorbitic specimen, polished before the deformation and not etched.

544. Pseudomorphic Persistence of the Old Austenite Grain Structure.— It is along its own grain boundaries that the austenite concentrates any finely divided mechanical impurities, slag, sulphides, etc., and it is here that they remain for instance during a quenching which results in the martensitic state, and a reheating to the sorbitic or granular pearlitic state. Hence these old boundaries still determine the path of least resistance along which the deformation of a polished surface is concentrated, as shown in Figs. B and D of Plate 27.

545. In martensite of medium carbon steel, deformation causes the typical martensitic markings to rise from a previously polished surface, quite like those which cooling in liquid air develops on a previously polished surface of austenite, and like the patterns which etching develops in martensite. Note in Fig. F of Plate 28 the nearly equilateral triangular figuring, such as the intersections of either octahedral cubic or 211 planes on an

[1] Ewing and Rosenhain showed like effects with the lead-tin eutectic, *Phil. Trans. Royal Soc.*, 1900, A, vol. 193, Figs. 47 and 48, after p. 375; *The Metallographist*, 1900, vol. 3, Figs. 40 and 41 on pp. 126 and 128.

octahedral face cause, at the left of *a*, and the arrow-head intersections at *b*, recalling the black martensitic arrow heads in Fig. A of Plate 2. In other fields these figurings are rhomboids. In still others they are rectangular, as cubic octahedral and 211 intersections of a cubic face are. These observations are compatible with the common belief that the martensitic structure is octahedral.

Under compression by shock the cracks in martensitic steel, though they may start parallel to the martensitic structure, yet often start and propagate wholly independently of it.[1]

The quenching cracks in martensite run in considerable part along the borders of the grains, which are probably the grains of the austenite of which the metal consisted at the moment of quenching. Yet the path of rupture is determined also in part by the intensity of the quenching stresses, which of course are irrelated to the grain boundaries, and hence in many cases it does not follow those boundaries[2] and indeed seems wholly capricious.

When a martensitic specimen of 0.40 carbon steel, the surface of which has previously been polished, has a crack propagated across it, I find that the surface deformation concentrates about that crack, as shown in Fig. E of Plate 28, in lines recalling the incipient slip bands of Fig. K of Plate 38.

The presence of a coarse but thin ferrite network in martensite does not appear (§753) to affect the fracture, according to Brinell's fracture studies. Hence we may infer that in such a composite mass rupture passes through the brittle martensite and avoids the ductile ferrite network, at least when this is narrow.

[1] Robin, *Rev. de Metallurgie, Mem.*, 1911, vol. 8, p. 453.

[2] Osmond, "Methode generale pour l'analyse micrographique des aciers au carbone," *Bull. Soc. Ind. d'Encouragement*, May, 1895. Also, "The Microscopic Analysis of Metals," Osmond, Stead, and Sidney, Griffin, London, 1913, p. 221.

CHAPTER 22

TWINS

546. Summary.—In the plastic deformation of metals certain crystalline blocks within a given grain instead of *slipping* may *twin*, that is their component crystal units may rotate into a new position symmetrical with their initial orientation. Twins, called also "macles," may be divided for convenience into the common "broad" or annealing twins and the narrow twins or "Neumann lamellæ" which appear on the surface as "Neumann bands." Annealing twins are very common in gamma iron or austenite, but are rare in alpha iron or ferrite, yet in this latter the Neumann lamellæ are developed easily by shock. Here they are "mechanical," that is, they form during the deformation itself, perhaps by "wheeling." Annealing twins, on the other hand, form not during the deformation but on annealing after it. They seem to form by "facing about."

While it is easy to see how slip and "wheeling" twinning may be the mechanism by means of which great plastic deformation is brought about, it is not easy to see how "facing-about" twinning, as it commonly occurs, can be.

A twin, called also a macle, is usually recognized in a polished metallic section as a parallel-sided area within a given grain, differing from the metal on either side of it in the directions of the slip bands, which zigzag on entering and leaving it, in the strength of the contrast between the treads and risers of the slip bands, in the tint to which it etches, and in the shape of its etching figures.

Twins are shown in Plates 29 and 30 as such parallel-sided areas, numbered 1, 2, 3, etc. The dark stripes that cross each of these areas are slip bands which zigzag as just indicated, but with one direction which is the same in all the odd-numbered areas, and with another direction which is the same for all the even-numbered ones, as in Figs. A, B, D and F of Plate 29, and Fig. F of Plate 30. In like manner, because Figs. C and D of Plate 30 represent oblique illumination, their alternate areas are uniformly bright or uniformly dark, as for instance in grain 1.

Strictly speaking, as each of two twin brothers is twin to the other, so each area is twin to its neighbor, as for instance in Fig. B of Plate 29, in which the odd-numbered areas are about as important as the even-numbered ones. But in many cases one set, for instance the even-numbered, may be much narrower than the odd, as in Figs. D, E and F of Plate 29, and E of Plate 30. Here the wider areas probably have the initial orientation, and the narrower ones the induced orientation. In these cases we speak with

perfect propriety of the narrow areas as "twinned areas" or "twins" and the wider ones as "the enclosing metal."

The zigzagging of the slip bands as they cross a series of twinned areas causes the effect of a staircase covered with a longitudinally striped carpet, treads alternating with risers, as shown characteristically in Figs. B and F of Plate 29. This is an optical illusion, for the surface is in fact plane, save for the slight transverse ribbing which each individual slip band represents. These apparent treads and risers are not to be confounded with the true risers which the slip bands themselves form between the true treads, which are the fragments of the initial polished surface between slip bands.

A twinned zone may not extend completely across the grain in which it occurs. This is the case with zone 2 of Fig. A of Plate 29, zones 3, 5, and 7 of Fig. B, and zone 4 of Fig. F of that plate. Again, a twin may change its width abruptly within a given grain, as zone 2 of Fig. E of Plate 29 does.

The difference in etching tint between the twinned zones and the enclosing metal is shown in Figs. C and D of Plate 30. These show also how, on rotating the specimen 90°, the relative brightness of the various zones is reversed.

Sections 547 to 552 consider twinning in native minerals in order to clarify the general subject; 553 to 556 take up the mode of occurrence of twinning in artificial metals and meteoric iron; and 557 to 561 the means of recognizing twinning. Sections 562 to 566 then ask whether what we call twins are really twins. In §567 I infer that the movement in plastic deformation does not occur in large part through twinning. Section 568 treats of the stability of twins, and §569 of their geometrical nature, as confirmatory evidence of the crystalline nature of metals.

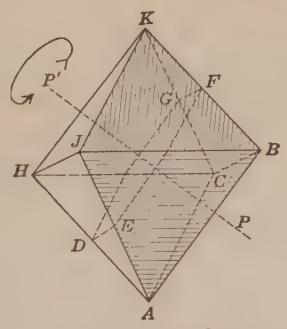

Fig. 90.—Simple octahedron.

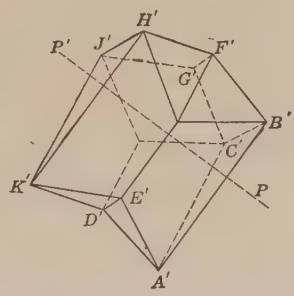

Fig. 91.—Hemitrope octahedron twinned by contact.[1]

547. Twinning in Idiomorphic Crystals.[2]—In the idiomorphic crystals of native minerals twinning is very common, both in its "juxtaposition" form shown in Fig. 91 and also in its "interpenetrating" form, in which two cubes, for instance, seem to have passed like two ghosts into and part way past each other, till their centers coincide, so that all six corners of each project out into the air.

If an octahedron, such as that of Fig. 90, is cut by a plane *DEFG* running through its center and parallel to two opposite faces *ABC* and *HJK*, and if

[1] After Watts's Dictionary of Chemistry, 1875, vol. 2, p. 160, Figs. 318 and 319.

[2] See article on "Crystallography," by L. J. Spencer, "Encyclopædia Britannica," 11th edition, 1910, vol. VII, p. 582.

its upper half is then rotated 180° about the axis PP' normal to that plane, the whole is thereby brought to the condition of a twin, Fig. 91. This simple case in which the twin consists of the two halves of a regular crystal thus rotated is called a "hemitrope."

We should not imagine that such a rotation of half a crystal about its axis actually occurs, but recognize that the relation between the two halves is that which such a rotation would cause.

Fig. 92 shows a vertical section taken through $A'K'P'P$ of Fig. 91. In the lower half of the crystal the outcrop or intersection on this section of the planes which are parallel to the octahedral face $A'D'E'$ of Fig. 91 is shown by the lines drawn parallel to AN. The rotation of the upper half of the crystal has of course carried its half of these octahedral planes with it, so that they are still parallel to the upper half of $F'G'J'H'$, and so that their intersections with the section shown in Fig. 92 are parallel to NK', and point west of north by the same degree by which those in the lower half point east of north. The two sets of intersections are shown by the inclined lines in Fig. 92. Note that the direction of these intersections changes abruptly at the junction line, and in the upper half of the crystal is symmetrical about that line with the direction of the corresponding ones in the lower half.

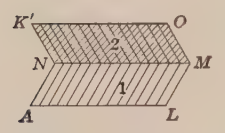

Fig. 92.—Imaginary herring bone structure in a twinned crystal.

548. A Simile.—It is somewhat as if a bricklayer, after laying his bricks pointing to the right as in the lower half of Fig. 93, on reaching a certain line AB, the line of junction or symmetry of the twins, began laying them pointing to the left as in the upper half, making a herring-bone pattern.[1] Thus we conceive that these little units fit into each other quite as closely along the line of twinning as in any other part of the pattern, so that this junction of the twinned areas lacks the triangular spaces or gussets, 6, 7, 8, and 9 of Fig. 65, p. 337, which arise at the grain boundaries, where the disregistering orientations of adjoining grains fail to match each other.

Fig. 93.—Imaginary herring-bone pattern in twinning.

Most twins in idiomorphic crystals have re-entering angles comparable with that at the left-hand end of the twinned octahedron of Fig. 91, and are thus distinguished from simple crystals. It is true that any two crystals growing out of each other may have a re-entering angle at their junction, but they are twins only in case they are symmetrical about the junction plane which passes through this angle.

549. Twinning and Parallel Growth in Idiomorphic Crystals.—The aggregations of crystals by parallel growth, like that of austenite shown in the frontispiece, while they resemble twins in having re-entering angles, differ from them in that the junction plane between the crystals thus aggregated is

[1] See Ewing, *Journ. Inst. Metals*, 1912, No. 2, vol. 8, p. 17. If the faces of these bricks are parallel to those of the octahedron, then the conception is slightly less simple, because in that case the bricks would not be rectangular but octahedral.

parallel to the edges of the crystals, whereas the junction plane between twins is parallel to one of the crystal faces, real or potential. The successive parts of a parallel-growth crystal have identical orientation, whereas the alternate parts of a twinned structure have not identical but symmetrical orientation.

Moreover, the cleavage and other structural planes are parallel throughout all the several crystals united to form a parallel-growth aggregation, but

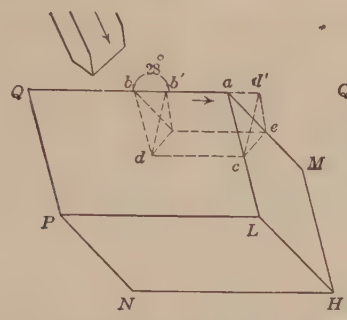

Fig. 94.—Forcible twinning of a calcite crystal about its glide plane.[1]

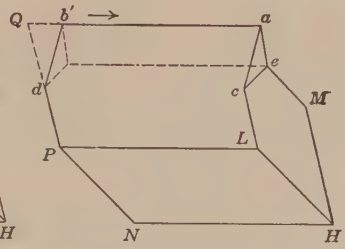

Fig. 95.—One secondary twin lamella forms.

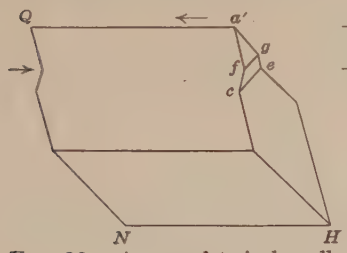

Fig. 96.—A second twin lamella forms.

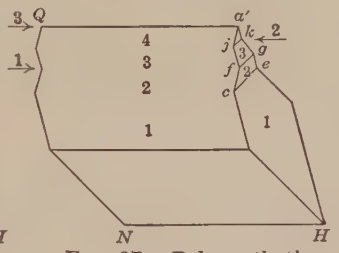

Fig. 97.—Polysynthetic twinning.

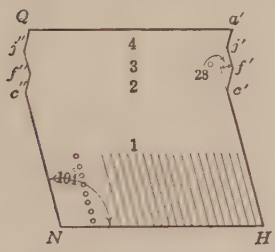

Fig. 98.—Plane section through $a'HNQ$ of Fig. 97.

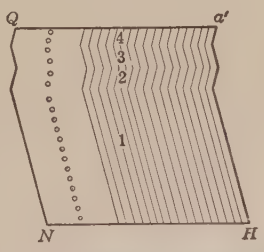

Fig. 99.—Section through $a'HNQ$ of Fig. 97, showing slip bands and etching figures.

in a twinned crystal they change at the junction plane $D'E'F'G'$, Fig. 91, so that their direction in the upper half is symmetrical about $D'E'F'G'$ with their direction in the lower half. And so with the optical properties, the etching figures, and indeed everything dependent on the internal structure.

These phenomena of twinning in idiomorphic crystals help us to understand the twinning of our artificial metals, though this is of a special type which we will now go on to consider.

[1] Baumhauer, "Encyclopædia Britannica," 1910, vol. VII, p. 585.

550. Artificial or Secondary Twinning of Native Minerals.—Baumhauer[1] showed that a crystal of the apparently perfectly brittle calcite could be twinned artificially, by inserting a knife-blade at the obtuse angle of a cleavage rhomb. In this way he rotated the edge *aecdb*, Fig. 94, from its initial position, shown in broken lines, in the direction of the arrow into the symmetrical position *a'ecdb'*. The surface *a'ec* was as bright after this rotation as when in its initial position.

Such artificial twinning, like every other form of post-genital twinning, is called "secondary."

In this rotation the obtuse solid angle at *a* became acute, *a'*, and the acute angle at *b* obtuse, *b'*. The plane *ecd* is called a "*gliding plane.*" It is also a "*twinning plane,*" because it is the plane about which the orientation of the corner *a'ecdb'* is symmetrical with that of the remainder of the crystal. Each little horizontal layer of crystal units has glided to the right to a new position in a single gliding plane parallel to *ecd*, though because all have moved simultaneously no one particle has glided past its neighbor. The movement in any slice parallel to the face *ecd* is nearly like that of the two platoons *AB* and *CD* of Fig. 103, p. 436, when they wheel about *B* and *D* as pivots into the twin positions *A'B* and *C'D*.

In thus gliding, any given particle has remained not only in this "gliding plane" but also in a plane parallel to one passing through *Qbb'aa'HN*, which is called a "plane of translation."

Such secondary twinning is very common in the fine grains of calcite in marble,[2] as shown in Fig. A of Plate 30. Here it is caused, as it is in metals, by plastic deformation under pressure. The black parallel-sided bands in each of the grains are twinned, that is, oriented symmetrically, with the lighter ones which enclose them.

551. The Twinning Orientation Symmetrical with the Initial.—The symmetrical arrangement of the brickwork of Fig. 93 and that of the two halves of the octahedron of Fig. 91 illustrate a universal law. For instance, the obtuse angle *c''NH* of the section cut through the calcite crystal of Fig. 95 on *Qa'HN* is shown in Fig. 98 to be 104°, or 14° more than a right angle. It is in conformity with this that the angle through which the corner of the crystal *abdce* of Fig. 94 rotated in twinning is $14° \times 2 = 28°$, as indicated also in Fig. 98. That is, the rotation inclines the plane of the face *bd* as far to the right of a perpendicular to *NH* as it was initially at its left. Thus the corner *a'b'dce* in its new position is part of a rhombohedron identical with the initial save that it is rotated 180° about a vertical axis, quite as the upper half of the octahedron of Fig. 91 is.

552. Polysynthetic or Multiple Twinning.—Because it is the pressure of the knife blade in the direction of the arrow of Fig. 94 that rocked the corner

[1] Groth, "Physikalische Krystallographie," 3d edition, 1895, Fig. 132, pp. 225, 226. "Like phenomena were examined by Mügge in crystals of antimony, bismuth, corundum, hematite, anhydrite, and diopside." *Idem*, p. 227.

[2] Adams and Nicolson, *Phil. Trans. Royal Society*, 1901, A, vol. 195, Fig. 1, Plate 24.

abdce into its twinned position, a like pressure fitly applied could rock the whole of the like corner of Fig. 95 into the like position *a'b'dce*. But the plane *cde* is evidently only one of numberless planes about which a like rocking might be caused. So, after this corner has been thus rocked as in Fig. 95, let a new stress in the direction of the arrow in Fig. 96 rock part of this same corner back from its present twinned direction into the initial direction inclined upward to the left, as here indicated.

Just as the first rocking in Fig. 95 broke the right-hand face of the crystal up into two planes intersecting at *ce*, so after this second rocking this face will be formed of three planes intersecting at *ce* and *fg*, the planes *a'fg* and *ceH* being parallel to each other and symmetrical with *cegf*.

Next a third stress along arrow 3 of Fig. 97 will cause a third slice of the corner to rock to the right, breaking the plane *a'fg* of Fig. 96 into the two conjugate planes *a'jk* and *jkgf*, so that the end face of the crystal is now made up of two groups of two planes each, with two re-entering angles, the even-numbered planes *inter se* and the odd ones *inter se* being parallel. These planes are separated by the parallel lines *ce*, *fg*, and *jk*, of which the first is the bottom of a slope, the second a ridge, and the third the bottom of a valley, while at the left of this last lies the peak *a'*. Proceeding in this way any number of parallel anticlinals and synclinals can in our imagination be set up on the end of the crystal, with striæ which are alternately ridges and valley-bottoms. Each ridge is connected with the valleys on either side of it by planes each of which is the end of a layer of the crystal lying between two planes parallel to *ced*. These layers are numbered 1, 2, 3, and 4 in Fig. 97. Each of them is twinned relatively to its immediate neighbors above and below. If they are thin enough we recognize on the end of the crystal only the striæ which represent the intersections of their end planes.

This multiple twinning, called "polysynthetic," is very common in metals. Its occurrence in the native minerals imbedded in a rock mass is illustrated in Fig. B of Plate 30.[1]

553. Twinning in Metals. Congenital, Mechanical, and Annealing Twins.—The twinning in metals which interests us is of the secondary as distinguished from the congenital type, and further it is usually of the polysynthetic or multiple type. Secondary twinning may be divided into the mechanical, caused directly by deformation, and the annealing, caused after deformation by heating highly enough to give mobility. Here the plastic deformation creates the twinning tendency, and the mobility given by the high temperature of the annealing enables this tendency to assert itself.

The zones numbered 1, 2, 3, etc., in Plate 29 are annealing twins; the narrow dark zones in Plate 32 are the mechanical twins in ferrite called "Neumann bands." Because annealing twins are usually much wider than the Neumann bands, these two classes usually differ greatly in their general appearance (§602).

[1] Polysynthetic twin striation of phagioclase in diabase. Iddings, "Rosenbusch's Rockmaking Minerals," New York, 1889, Plate 25, Fig. 6.

This chapter considers annealing twins primarily, the following one taking up the Neumann bands, and Chapter 24 the mechanism of twinning.

554. Congenital Twinning.—Twinning might well be supposed to occur congenitally in a large proportion of cases, that is, during the actual passage from the molten to the solid state, from the fact that the twinned idiomorphic crystals of native minerals, growing freely in cavities and presumably deposited from aqueous solutions in many cases, are naturally supposed to have twinned congenitally, because we do not readily conceive of any post-genital plastic deformation which could cause this often perfect twinning, this rotation of long thin half-crystals from an initial simple form. That the twinned crystal of gypsum shown in Fig. 99*A* could have been formed by an external force which rotated half of it through 180° into its present shape is not readily believed. Osmond and Cartaud[1] seem to have thought that twinning in metals is congenital in a not unimportant fraction of the cases. Professor William Campbell has beautiful congenital twins of copper deposited electrolytically from solution.

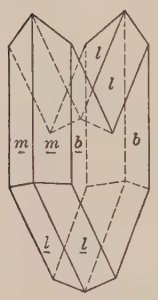

FIG. 99*A*.—Twinned crystal of gypsum.[2]

But apart from such electro-deposition of idiomorphic crystals, I have seen no industrial cases in which we can be sure that the twinning is congenital and not mechanical or annealing, because the deformation needed to cause annealing twinning, and in the case of tin to cause mechanical twinning, is so very slight that it may always be suspected as the real cause unless great precautions are used. Idiomorphic twins, of course, need not be under this suspicion.

The curved twin shown in a crystal of native copper in Figs. AA and AB of Plate 30 is probably congenital, because there has been no heating to cause an annealing twin, and because the very slight deformation given this crystal could hardly cause a mechanical one.

555. The mechanical twinning of metals, though it has been questioned,[3] is very familiar in the form of the Neumann lamellæ of iron, discussed in the next chapter. It occurs in tin (§609) and is probably common in other metals.

556. Annealing Twinning.—In the very great majority of cases the artificial twinning of metals is due to annealing after deformation. We may conceive that the plastic deformation throws great stress on certain rows of crystal units, tending to cause them to rotate into the twinned position, but that this tendency does not take effect, does not actually cause rotation, till the high temperature of annealing gives the mobility which permits obedience to this stress.

Rosenhain and Ewen find that even the slight deformation caused by the saw cut for removing the specimen from the rest of the cast mass,

[1] *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, pp. 444, 488.
[2] "Encyclopædia Britannica," 1910, vol. VII, p. 582, Fig. 81.
[3] Rosenhain, *Engineering*, 1913, vol. 96, p. 510.

or by blowholes, or by polishing, may suffice to cause twinning on later annealing.[1]

Annealing twinning is caused easily, on a scale and in a form readily recognized, in many metals and alloys,[2] for instance gold, silver, copper, brass, tin,[3] lead, and austenite or gamma iron, and not only in that of manganese steel and invar of Plate 29, but also in that of carbon steel when above the transformation range.[4]

Annealing twins, and in general those of their order of width, are so rare in alpha iron or ferrite that I should suspect that the few which have been reported are simply unusually wide mechanical twins, were it not that some broad twins which I have found seem clearly to be annealing ones.[5] Ground for such suspicion is given by the width of the twins of Fig. F of Plate 30. Though these twins are of the order of width of annealing twins, they are almost certainly mechanical ones. (§600, footnote.)[6]

557. Recognizing Twinning in Metallic Sections.—In rock masses the twinning of their component minerals is shown (1) by the presence of re-entering angles, (2) by the optical properties, (3) in etched sections by the presence of parallel-sided bands differing in tint from the adjoining surface, and (4) by the deflection or zigzagging of rows of etching figures. Even here the re-entering angles can be observed only in certain forms of twinning and only when the enclosed crystals are idiomorphic, as they are only very rarely in metals. Moreover, the optical phenomena do not apply to crystals of the isometric system, to which most metals belong, and the usual mode of observing them, by transmitted light, is excluded by the opaqueness of the metals themselves. The first two methods thus being inapplicable to me-

[1] "Intercrystalline Cohesion in Metals," *Journ. Inst. Metals*, 1912, No. 2, vol. 8, p. 171. Of two like silver castings from the same melt, one which was polished before vacuum etching was found markedly twinned on etching, whereas the other, exactly like it save that the polishing was omitted, was not twinned at all. Here the polishing evidently caused enough deformation to determine twinning on heating.

[2] *Journ. Inst. Metals*, 1912, No. 2, vol. 8, p. 16; and *Proc. Staffordshire Iron and Steel Inst.*, 1910, vol. 26, p. 32.

[3] For striking micrographs of twins in brass, tin, and copper, aluminum alloys see Edwards and Carpenter, "The Hardening of Metals," *Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, Plates XIII and XIV, following p. 176. Rosenhain and Ewen give an extraordinarily beautiful micrograph of twinned silver on Plate XXV, *Journ. Inst. Metals*, 1912, No. 2, vol. 8, after p. 168 (Rosenhain, "Physical Metallurgy," 1914, Fig. 107, opp. p. 251).

[4] See Baykoff, *Rev. de Met.*, 1909, vol. 6, p. 832. His Fig. 10, the structure of steel of 1.94 carbon and traces of manganese, etched with hydrochloric acid at 1120°, shows very clearly marked twinnings which do not run completely across the grains in which they occur.

[5] Mügge found that ferrite might twin. Rosenhain finds cases of the usual broad form of twinning in transformer iron, or very pure ferrite, strained in tension not far beyond the elastic limit (*Journ. Iron and Steel Inst.*, 1904, No. I, vol. 65, p. 354, and Fig. 15). Heyn in 1904 had never found a case in undeformed iron, and at most one which he believed to be such a case in deformed iron (*idem*, p. 377). In his "Materialenkunde" (Martens-Heyn, II, A, 1912, p. 221 and Fig. 49, Plate 10) he illustrates what he seems to believe a possible twin in ferrite. I find that broad mechanical twins can be developed abundantly in both the A.R.M. steel and the thermit steel, when coarse.

[6] For the use of this thermit product I have to thank Professor William Campbell, and through him the Goldschmidt Thermit Company of New York.

tallic sections, we rely on the last two and more especially on one closely related to the last, the behavior of slip bands on entering and leaving a twinned area. Let us now consider these.

558. Zigzagging of Slip Bands as Evidence of Twinning.—A section through $Qa'HN$ of Fig. 97 would be like Fig. 98. Here you will have to force your imagination to recognize that this figure and Fig. 99 show true planes and not stepped surfaces. To aid in this effort I indicate in Fig. 98 only the beginning of the row of etching pits and of the slip bands, so that the planeness may here be recognized, while in Fig. 99, which represents the same plane, I continue the pits and bands across the figure. The horizontal zones 1, 2, 3. and 4 are sections of the several layers 1, 2, 3, and 4 of Fig. 97. The line ce, the ridge fg, the valley jk, and the peak a' are represented in this section by the sawteeth $Hc'f'j'a'$.

If the striæ in the lower part of layer 1, Fig. 98, represent the slip-plane outcrops, that is, the intersection of the plane of the paper with the slip planes parallel to the end ceH of layer 1 of Fig. 97, and if we extend them upward across the section as in Fig. 99, then because rotating layer 2 into a position symmetrical with that of layer 1 has moved the line $c'f'$ of Fig. 98 into a position symmetrical with that of Hc', and because these intersections must be parallel to $c'f'$ which is itself only the last of them, they must here turn sharply to the right, making the angle $Hc'f'$.

For like reason they must deflect to the left on reaching layer 3 and back again to the right on reaching layer 4. But recognize that it is only an optical illusion that now makes Fig. 99 look like a set of steps. It is the same plane as that of Fig. 98. It is our misfortune that our minds are so constituted that they insist rather stubbornly on misinterpreting such a hatched figure as a stairway.

This zigzagging of the crystalline planes naturally leads to a corresponding zigzagging of the slip bands, as shown in Figs. A, B, E, and F of Plate 29. This zigzagging is the usual symptom by which we recognize twinning.

Or, if the row of circles in Fig. 98 is a line of etching pits following along some crystalline plane which is more easily corroded than the rest of the surface, then if, as in Fig. 99, it is continued up to the top of the section, it too must zigzag on entering each new layer 2, 3, and 4.

Such a zigzagging of a line of etching pits into a direction symmetrical with its initial direction and back again, and so on back and forth, can be identified at times in the crystals of some native minerals as conforming with the re-entering angles and with the striæ on the ends of the crystals which testify to twinning. Because this symmetrical zigzagging of the etching pits is a natural consequence of the change of orientation back and forth from twinned area to area, and because it does not connect itself with any other cause, it is taken as an indication of twinning.

The contrast between the treads and risers of the slip bands is sometimes about the same in the odd as in the even-numbered zones, as in Figs. C

and F of Plate 29 and in parts of Fig. B of Plate 29, but more often it is stronger in one than in the other. Proceeding to the extreme, the slip bands so prominent in zones 1 and 3 of Fig. D of Plate 29 are hardly visible in zone 2. A like eclipsing occurs in parts of the twin of Fig. AA, and of twins 2 and 6 of Fig. F of Plate 30. Between these two extremes, the equality of contrast of Figs. C and F of Plate 29 and the eclipsing in these last cases, are the twins of Figs. A and E of Plate 29. In these, moreover, the slip bands follow two conjugate directions which are common to the odd and the even-numbered zones, but with a change from zone to zone in their relative prominence. The zigzagging of a single set of slip bands may be regarded as an extreme case of this type, the prominence of one set becoming zero in the even-numbered zones and that of the other set in the odd-numbered zones, so that there seems to be only one set which zigzags.

The angle of deflection of the slip bands on entering a twinned zone must, of course, vary greatly with the inclination of the plane of the microsection to that of the twin. It is about 45° in Fig. E and much less in Fig. C of Plate 29. In Fig. AA of Plate 30 it is extremely small. It must occasionally happen that the angle is no greater than the irregularities in the slip bands themselves, in which case the slip bands would cross the twinned zone without apparent zigzagging.

To sum up the behavior of slip bands at a twinned zone, they may be eclipsed; or if they continue into it they usually change direction, and they often change the strength of contrast between their steps and risers, with the result that if there are two conjugate sets their relative prominence is reversed. This change of relative prominence, if pushed so far as to efface the less prominent set, would give rise to the usual case of a single zigzagging set.

559. Sequence of Operations Needed to Disclose Twinning by Means of Slip Bands.—In order to disclose twinning by means of zigzagging slip bands, the twinning must already exist when the slip occurs. Further, the slip bands can be seen only on a smooth surface. Hence the usual order of operations in causing these zigzagged slip bands is: (1) plastic deformation to create the twinning tendency; (2) annealing to enable the twinning tendency to assert itself; (3) polishing to enable us to see the slip bands which are later to demonstrate the existence of the twinning; (4) a second plastic deformation to form the slip bands.

560. Variations in the Regularity of Zigzagging.—The regularity and conspicuousness of the zigzagging vary greatly, from the degree shown in Fig. F of Plate 29, through that in Figs. A, D, and E, of that plate to that in Fig. C. In the staircase of Fig. F, each tread is one twinned zone, and each riser a zone symmetrical with the treads to right and left of it. In Fig. D a narrow shelf, 2, in which the faint slip bands run at about 45°, looks as if it were an offset in a steep cliff, on which the slip bands run about N-80°-W.

The essential characteristics are: first, that in each tread and in each

riser the slip bands are parallel; second, that their direction in each tread is the same as in every other tread, and that they have a different but common direction in all the risers; and third, that the intersections of the various treads and risers are approximately parallel.

561. Parallel-sided Zones as Evidence of Twinning (See §558).—Under oblique illumination a twinned layer in an etched section differs in brightness from its neighbors because its facets are differently inclined. This is true, for instance, of the black and white stripes of grain 1 of Figs. C and D of Plate 30, showing the same field with different directions of oblique illumination, the parts which are white in C being black in D and *vice versa* (see §380). Moreover, the sides of twinned zones are usually straight and parallel, or made up of parallel fractions as in Plates 29 and 30. Such parallel-sided areas differing in brightness from their neighbors are usually twinned areas.[1] Thus among the symptoms of twinning the presence of a straight and parallel-sided area is more constant than a change either in the direction of the slip bands or in the contrast between their treads and risers. A straight parallel-sided area which changes either the direction of the slip bands, or the contrast between their treads and risers, or extinguishes the slip bands, or changes the relative prominence of two conjugate sets of them, is to be interpreted as a twin, and those which persistently do none of these things are not.

562. Are the Parallel-sided Zones in Metals Truly Twins?—I refer here to these zones which are habitually if not universally spoken of as twins. Let us ask whether they really are twins.

Their orientation is clearly different from that of the enclosing metal, because they differ from it in the tint which they assume on etching or heating, and because the slip bands change direction on entering and leaving them.

In such a staircase structure as that of Figs. B or F of Plate 29, all the treads evidently have a common orientation, and all the risers another common orientation. Each tread or riser is the equivalent of a zone or lamella. All the odd-numbered zones have one common orientation, and all the even-numbered ones another. We assume that these are twins; but do we know this? If they are truly twins, the difference must be such that the orientation of the even-numbered zones is symmetrical crystallographically with that of the odd-numbered ones. In the case of transparent minerals the question whether this symmetry between successive zones exists can be answered positively by determining their optical properties, and the recent application of the X-rays to determining the crystallography of crys-

[1] For an interesting case in which a pseudo-twin is proved not to be a true twin by the fact that the slip bands cross it without deflection, see Rosenhain, *Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, Plate XVIII, Fig. B, opposite p. 180; Plate XIX, Fig. C; and also p. 181. In this case the area differs in brightness from the rest of the field quite as a twinned area does and in general suggests twinning, except first in having one side strongly curved, second in not deflecting the slip bands, and third that on polishing and re-etching it ceases to differ in brightness from the rest of the field.

talline metallic surfaces raises hopes that we may be able to answer this question for metals also.

In transparent minerals such a zonal arrangement does not always represent twinning in this legitimate sense, just as re-entering angles on the surface of groups of crystals, though they may suggest twinning, are no proof of it.

563. Their Dynamic Origin Suggests that They are Twins.—A strong suggestion, though not a proof, of this symmetrical relation between the alternate zones in metals comes from their dynamic origin. Just as when a crystal of calcite is cleft as shown in Fig. 94, the rotation of the corner ceases naturally when it has reached a position symmetrical with the rest of the crystal, so the rotation of the particles of a metal brought about by mechanical means, whether or not the rotation needs the aid of annealing, suggests that the position on reaching which this rotation ceases, the natural end of the rotation, is symmetrical with the initial. This is all the more true because all these strata rotate through the same degree, as is shown by their having uniform orientation. The fact that they all rotate through the same angle certainly suggests that this angle is a crystallographic one, related to the internal symmetry.

A condition which, in certain minerals, may lead to a zonal system without symmetry, is lacking here. Such a zonal system may arise from the development of crystalline growth from two independent and irrelated origins. Because each of these two growths adheres to its own initial crystalline orientation and because the two initial orientations are irrelated, the alternate zones are irrelated and hence not symmetrical. There is no reason why they should be.

But the alternate zones created in metals by secondary twinning were initially one, with common orientation. It is natural to infer that the new orientation into which the odd-numbered zones have rotated is related to the initial orientation which it has replaced. If so, it must be identically related to that of the even-numbered zones. Of such relations the symmetrical or twinned one would seem the most natural.

564. Their Conjugate Arrangement Suggests Twinning.—Though the more prominent of the slip bands in zones 1 and 3 of Fig. E of Plate 29 differ in inclination from the more prominent ones in zone 2, yet in each set of areas there is a second and less prominent set of slip bands parallel to the more prominent slip bands in the other set of zones, so that the direction of the slip bands is really the same in all three, and the difference is that the horizontal slip bands are more prominent than the N-45°-E ones in zone 2, whereas the converse is true of zones 1 and 3.

If, to fix our ideas, we assume that the horizontal slip bands of zone 2 represent the planes of easiest slip and the inclined ones the planes next in ease of slip, the identity of direction in the odd and even-numbered areas suggests that the rotation has been such as to bring the planes of easiest slip in the twin parallel to those next in ease in the enclosing metal, and *vice versa*. Such a relation certainly suggests symmetry very strongly.

565. That They do not Affect the Hardness Suggests that They are Twins.—The hardness of metals is affected so readily by structural changes that the fact that removing this zonal arrangement does not affect the hardness, in the case of the Neumann strata as reported by Stead,[1] gives an additional indication that the odd and the even-numbered zones have symmetrical orientation, and therefore dovetail into each other as if their orientation were uniform. If the two orientations, in differing, were not symmetrical, they could not thus dovetail into each other, and would therefore naturally be expected to give rise to irregular gusset spaces which, as in the case of slipped regions, would necessarily be amorphous, and hence would be accompanied by an increase of hardness and by the other physical changes which slip causes. That twinning need not cause amorphization is explained in §619.

The reasoning which is given at the end of §598 tending to show that the narrow twins called Neumann lamellæ are true twins, tends by analogy to show that the same is true of the usual broad annealing twins.

These considerations, though they do not prove rigorously that these alternate zones are true twins, at least make it so probable antecedently that it may well be taken as a working hypothesis.

566. Inconspicuous Twinning.—The characteristics of twins can be traced in the left-hand part of Fig. A of Plate 24. What at first looks like a random curving of the slip bands on examination is seen to consist of a zigzagging back and forth between two fixed directions. The slip bands in the areas *aa'*, *cc'*, *ee'* point about N-40°-W, whereas those in the alternate areas *bb'*, *dd'*, and *ff'* point about N-10°-W. The areas *b*, *d*, *e*, and *f* are plain enough; but the intermediate one *cc'* is an important link in connecting the more regular with the less regular cases of twinning.

Still less conspicuous is the zigzagging in manganese steel shown in Fig. C, Plate 29. On raising the book so that the plane of the paper is nearly in the line of sight, we see a band running from *a* to *b* in crossing which each of the strong N-25°-E lines zigzags slightly. We next recognize a band *cd* parallel to *ab* in which a like zigzagging occurs. Looking back at Fig. A of Plate 24 we see that the phenomena are essentially the same in both, save that in Fig. A of Plate 24 there are more reversals, and these are more pronounced, than in Fig. C of Plate 29. The latter are only the last term of the series beginning in Fig. F of Plate 29. In every case there is a series of roughly parallel treads alternating with roughly parallel risers, with roughly parallel intersections, carpeted lengthwise of the staircase with a longitudinally striped carpet. In Fig. C of Plate 29 two parallel areas, *ab* and *cd*, cross the N-25°-E lines obliquely, deflecting all these lines repeatedly from one to the other of two conjugate directions, but without lessening their continuity, strength, or parallelism in the least, and thus having inconspicuously all the essential properties of twinned areas. They might be referred to slight inequalities in the surface of the section were it not that such inequalities would

[1] *Trans. Faraday Soc.*, 1915, vol. 10, p. 276.

throw parts of the N-25°-E lines out of focus under this great magnification, 500 diameters.

The narrowness of these twins, reminding us of the narrow Neumann bands, is noteworthy.

567. Annealing Twinning does not Disturb Previously Polished Surfaces.—In order to learn whether it does, I had a truncated wedge of pure Lake Superior copper deformed, so as to set up the twinning tendency, polished, and heated to 500° for 1 hour in a sealed quartz tube. This left the polished surface as bright and smooth as before, without a sign of deformation, or any marking detectable either by the naked eye or with the microscope. Yet during the heating the metal had twinned abundantly, as was next proved by deforming it so as to cause slip bands, which zigzagged in the usual way on entering and leaving the simultaneously formed parallel-sided areas, now so familiar as twinned areas. Thus, though during the heating the metal had twinned abundantly, this twinning had not disturbed its surface in the least.

It is true that twinning figures may form on such a previously polished surface on annealing *in vacuo*, or even in the air in the case of metals which do not oxidize in the air, such as silver and platinum. This may be due in part to the volatilization of the surface layer made amorphous by the polishing, thus revealing the twinning of the underlying crystalline metal, but it is evidently due also in part to the spreading of the crystallization of the underlying metal up through the surface layer. That this last assertion is true is shown by the fact that etched surfaces, which because etched have a configuration representing the crystalline structure at the time of etching, if now reheated *in vacuo* in such a way as to change the crystalline structure of the underlying metal, may themselves change and adopt a configuration representing this new structure.[1]

But this exhibition of twinning figures on heating a polished surface neither indicates appreciable motion nor detracts from the force of my case described above, in which the heating, though causing twinning, caused no motion of the previously polished surface.

568. The Stability of Twins.—If what we take for twins are true twins, one sees no reason why they should be removed by heating, unless this is accompanied either by a transformation, such as that from ferrite to austenite, or by grain growth, either of which processes might well efface them. There is no reason why a grain which is forced, by grain growth, to lose its initial orientation and adopt a new one, should remain twinned through this change. Edwards and Carpenter[2] find that twins do disappear

[1] Rosenhain and Ewen, "Intercrystalline Cohesion in Metals," *Journ. Inst. Metals*, 1912, No. 2, vol. 8, Figs. 8 and 9, after p. 168.

[2] "The Hardening of Metals," *Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, p. 148. It is true that Rosenhain objects to this inference (*idem*, p. 181, and "Metals, Amorphous and Crystalline," *Engineering*, 1913, vol. 96, p. 509), on the ground that what these investigators take for twins are not. But the evidence since given by Edwards that these are twins ("Metallic Crystal Twinning by Direct Mechanical Strain," *Journ. Inst. Metals*, 1915, No. 2, to appear) seems to meet this objection.

in certain metals on heating, but they do not indicate whether this disappearance can be explained by simultaneous grain growth. The disappearance of twins from brass of 70 per cent. copper and 30 per cent. zinc on heating to 850°, noted by Stead, is certainly referable to the rapid grain growth which would occur at a temperature so slightly below the solidus of this alloy.[1]

My preliminary experiments with copper and invar have detected no tendency on the part of twins to disappear on heating to 950°.[2]

The thin twins called Neumann lamellæ certainly disappear on heating ferrite to above the transformation range, that is to above *Ac*3 (§585). But this is a natural result of the complete double transformation, from ferrite into austenite in heating and back again in cooling. The metamorphic power of the transformation is familiar through its refining the grain size, its giving and removing the hardening power, its removing and restoring the magnetic properties, and other striking changes.

569. Twinning is geometrical, indeed strikingly geometrical, because it arrests rotation when it has reached a position symmetrical with the original. This shows that the geometrical internal structure of the crystal represents some agency the potency of which did not cease to exist when it had completed the crystal, but persist and govern the motions of the little blocks which compose the crystal when these are forced by external agencies out of the position in which the crystalline force had set them initially.

[1] *Journ. West of Scotland Iron and Steel Inst.*, 1912, vol. 19, p. 189.

[2] Specimens of copper and of invar were deformed plastically so as to develop the twinning tendency; annealed so as to cause twinning; and polished and etched. They then showed abundant twinning even under direct illumination. One specimen of each metal was then held at 800° for 2 hours, and another at 950° for 2 hours. On repolishing and etching the specimens the twins were as conspicuous as before. A specimen of invar which had been exposed to 800° was deformed so as to cause slip bands anew. These zigzagged as before, marking out broad twinned areas like those formed before this heating. This proves that twins can persist through an exposure of 2 hours to 950° in both these substances.

CHAPTER 23

THE NEUMANN BANDS OR MECHANICAL TWINS IN FERRITE

570. Summary.—The Neumann bands,[1] commonly called Neumann "lines"[2] are the outcrops of thin mechanically twinned lamellæ or strata caused in ferrite and some other metals[3] by deformation, and most readily by shock. In ferrite they are destroyed by heating to above *Ac*3. On smooth surfaces, either those of fractures, or those polished before the deformation, they are often easily visible to the naked eye, as in the case of those in Figs. C and D of Plate 35. They are disclosed readily by etching. Their axes are usually straight, and so are their edges on unetched surfaces, though often serrated and otherwise irregular on etched ones.

The lamellæ are usually parallel to the faces of the {211} trapezohedron, Fig. 37, but probably follow the cubic cleavages occasionally. The number of possible 211 and cubic directions is 6 + 2 on a cube face, 9 on an octahedral face, 5 + 1 on a rhombo-dodecahedral face, and might reach 12 + 3 on a random face.

On a section parallel to a cubic face these directions run either diagonally of the square which constitutes that face or else from its corners to the middle of the opposite sides, and hence make angles of 26° 34′ and 63° 26′ with these sides. Those which represent the cubic cleavages are, of course, parallel to the sides of the cube face.

The Neumann bands differ from slip bands in being of a higher order of magnitude, and much stiffer and straighter in ferrite, and probably in reappearing always on repolishing and re-etching even after moderate heating. They are narrower, and usually much narrower, than annealing twins, from which they may differ in representing wheeling instead of facing about.

If slip bands are generated in a field in which Neumann bands are already present, at least in ferrite and chrome-nickel austenite, one set of the slip bands runs parallel to or alongside of them, as shown in Fig. 102.

They were discovered in meteoric iron in 1850 by Neumann, for whom they were named, and later in artificial iron by Prestel.

[1] Compare Cohen, "Meteoritenkunde," Stuttgart, 1894, and Osmond and Cartaud, "The Crystallography of Iron," *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, p. 462.

[2] To call them "lines" is inaccurate, because their thickness is always appreciable, and sometimes very considerable, as in Figs. F of Plate 30 and D of Plate 32.

[3] The narrow bands in the bismuth of Fig. D of Plate 41 are Neumann bands. Note that at *NOP* at the left of this figure they deflect in the staircase way in which slip bands zigzag in crossing a twinned lamella, so that here we have twins of two orders of magnitude, the Neumann lamellæ, and the broad bands such as *O*. Compare Desch, "Note on Crystal Twinning and the Martensitic Structure," *Trans. Faraday Soc.*, 1915, vol. 10, p. 255.

In §§571 to 583 and 586 I consider their general characteristics, and in §§584 and 585 the conditions under which they form and disappear. In §579 I show that they are steps. In §588 I show how they can be distinguished from slip bands. Sections 589 to 602 consider their crystallography. Section 603 shows that they may direct the path of rupture; §§604 and 605 show their behavior at grain boundaries and on intersecting each other; and §606 shows that they do not affect the hardness of the mass greatly if at all. The mechanism of their formation is considered in the following chapter.

Let us now consider these matters in greater detail.

571. Their General Appearance.—In fractures they form narrow straight bands, running in definite geometrical directions, black under vertical light, as in Fig. B of Plate 35, but bright under oblique light, as in Figs. C and D. In Fig. B the broad black band *G* is probably a Neumann band, and so are the narrower ones making an angle of 45° with it, and all the bright bands in Figs. C and D.

When developed by deformation on a previously polished and unetched surface they form nearly straight narrow lines, often black but occasionally light under favorable illumination, as in the upper left hand part of Fig. C of Plate 32, and as at *A* to *F* in Fig. D.

Cases of slight long-radius curvature are considered in §593.

572. Their Appearance on Etched Sections (§576).—Etching usually gives a very false impression of their true size and shape, by widening them not only greatly but irregularly, so that their irregularity in etched sections may be as striking as their regularity in unetched ones, as is seen on comparing the etched sections of Plates 33 to 36 with the unetched ones of Plate 32.[1] A spotty[2] widening is seen in Figs. A and E of Plate 34. In the former of these, which represents an unpolished fracture, the Neumann bands themselves are really so narrow as to be invisible here between the spots of any one row, yet the circles and continuous bands formed by etching are about ⅛ in. wide even on this slight magnification. The constancy of this false width is striking. Evidence supporting the inference that these individual Neumann bands are extremely narrow is given in §620. Etching in other cases develops serrations, as in Fig. E of Plate 33 and D of Plate 34, and leads to very varying degrees of regularity, as shown by the other figures of Plates 33 and 34, and by Figs. E, F, and G of Plate 35.

Their irregularity, which at one time led Osmond[3] to hold that some at least of them are not twins in the usual sense, seems to me an etching effect,

[1] The etching was done with picric acid for all of Plate 33, Figs. B and D of Plate 34, and A to F, H, and I of Plate 36; first with 10 per cent. and then with 2 per cent. aqueous nitric acid for Figs. A and F of Plate 34, and E, F, and G of Plate 35, and with Heyn's ammonio-cupric-chloride-water reagent for Fig. E of Plate 34.

[2] Heyn's reagent tends to cause this spottiness, but picric and nitric acid tend rather to cause furrows, though in this silicon steel of Fig. A of Plate 34 picric acid causes these round spots, running together into bands.

[3] Osmond, Fremont, and Cartaud, *Rev. Metallurgie, Mem.*, 1904, vol. 1, p. 24.

for on unetched sections they are of the order of regularity of the annealing twins, as is seen by comparing the unetched Neumann bands of Plates 31 and 32 with the annealing twins of Plates 29 and 30.

In spite of their irregularity, etched Neumann bands generally have straight axes, and suggest stiffness when compared with slip bands, which often suggest great flexibility (§498 *et seq.*).

On etching a polished or cleavage surface the Neumann bands may appear first as disconnected spots, such as are shown in Fig. A of Plate 34, which later increase in number and run together into regular bands. An early stage of this process is shown in the band R parallel to C, a later one at D, in which disconnected circles and intersecting ones occur in a band of which the greater part is continuous.

These bands are developed well so as to show their width by etching with either alcoholic picric acid or very dilute nitric acid (2:100). More severe etching, as with 1:10 nitric acid for ½ minute after 4 minutes with copper ammonium chloride, enlarges them greatly and channels them. Bas-relief polishing shows them as slight depressions on a cubic face, but on other faces sometimes as depressions and sometimes in relief. They may then look like fine file teeth, and if abundant enough may give the specimen a watered look.[1]

573. Possible Neumann Bands.—The band MM' of Fig. E of Plate 33, especially because of its serrations, looks like a Neumann band greatly widened by etching, and NN' also might well be taken for one. Yet doubt is thrown on their nature by their failure to zigzag the apparent slip bands a to e. The reason for interpreting these as slip bands is that they were formed by quiescent deformation which seems incompetent to cause Neumann bands in this A.R.M. steel. Moreover, they are too narrow to represent Neumann bands on this magnification. If not Neumann bands they should be slip bands.

The interpretation of this figure needs further study. It is possible that the Neumann lamella itself is so narrow, in spite of the width of the etching band about it, that the deflection which it causes in these slip bands escapes detection.

574. They Persist on Deep Etching.—These bands persist not only on deep etching but also on repolishing and re-etching, a characteristic which often serves to distinguish them from roughly similar bands due to scratches. A scratch may harden the metal for an appreciable distance below the surface. If it is now flowed over by the polishing, and the specimen is again deformed for the purpose of developing slip bands or Neumann bands, the hardened layer may behave differently from the enclosing metal, both in the deformation and in the following etching, and may thus give rise to something

[1] The assertions contained in this paragraph are based on the observations of Osmond and Cartaud, *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, p. 463. For very deep etching of these Neumann bands Portevin and Durand use successively 2 per cent. alcoholic nitric acid for 4 hours, Heyn's reagent for 3 minutes, and 10 per cent. nitric acid for 30 minutes (*Rev. de Metallurgie, Mem.*, 1914, vol. 11, p. 774).

much like a Neumann band. Deep etching erases a scratch by passing beyond this thin hardened layer, but not beyond a Neumann lamella.

575. A nodal etching effect along the Neumann bands is shown in Fig. D of Plate 33, and much more strikingly in Fig. B of Plate 34, both of the A.R.M. steel. Starting at the bottom of this figure, the first three Neumann bands, *K*, *IJ*, and *GH*, are pretty regular. Then comes the less regular *EF*, with indentations recalling those of Fig. D of Plate 34, then *CD* with like indentations at its right-hand end, turning into a nodal arrangement at its left-hand end, and then the strikingly nodal *AB*. Even without this gradual progressive development of the nodality from bottom to top of this figure, the exact parallelism of the upper bands, *AB* and *CD*, to the lower ones which are clearly Neumann bands, would leave no doubt that the former also are. Note that at the left-hand end of *CD* the band is broken up into disconnected islands.

The way in which two nodes seem to evade each other, as shown in Fig. H of Plate 36, is interesting.

The etching circles of Fig. A of Plate 34 seem to represent another type of etching. Still a third is given by the banded etching of Fig. I of Plate 36, which is a detail of one of the apparent Neumann bands of Fig. F, Plate 36.

The explanation of this etching figuring remains to be studied. In most cases a special reason for not classing masses thus arranged along the 211 planes as impurities is that these are habitually arranged about the octahedral Widmanstättian planes of the austenite. This reason may not apply to this case, because the absence of the *A*3 retardation in this silicon steel may mean that the metal is ferrite even up to relatively very high temperatures, and hence has the 211 structure, and hence in fine distributes foreign matter about the 211 planes at the high temperatures of great mobility, whether during that distribution the temperature is rising or stationary.

576. The serration in Fig. C of Plate 33 in the etched Neumann bands *M* and *N* is complicated by what looks like faulting by the slip caused by squeezing the specimen, after causing these Neumann bands in it by hammering, and then again polishing it, so that the slip bands might become visible. This squeezing cannot cause Neumann bands in this material. Hence these ripple marks running horizontally in this figure should be slip bands.

Osmond and Cartaud[1] inferred that these serrations were probably connected with a system of Neumann lamellæ following a different direction. This inference is supported by Fig. E of Plate 33. Of the re-entrant serration *fg*, the side *f* is parallel to line *F*, and *g* to lines *a*, *b*, *c*, *d*, and *e*. For our present purpose it is immaterial whether these lines are slip bands or Neumann bands, because these two kinds of bands have common crystallographic directions, so that the direction of a slip band which is seen may well be that of Neumann bands which are not seen. Or this con-

[1] *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, p. 463.

nection may be with a set of 211 planes rather than with Neumann lamellæ actually developed along those planes.

577. The furrows formed along the Neumann bands by etching them, as shown in Fig. F of Plate 34, are seen in cross-section, as at 1, 2, and 3 in Fig. E and in Fig. F of Plate 36, to be flat and straight sided, but of variable inclination and apparently not symmetrical. These figures represent a cross-section of the specimen of Fig. E of Plate 35 after the grooves there shown had been etched in it. This section was cut diagonally of the specimen and therefore parallel to one set of etching grooves and at right angles to the other. It tends to confirm the impression of irregularity in these grooves which Fig. D as well as Figs. E, F, and G of Plate 35 give. Thus in Fig. E of Plate 36 groove 1 is flatter than 2, and 2 than 3. Moreover neither 2 nor 3 seems to be symmetrical.

The observed internal angles of these furrows as seen in this section vary from 127° to 146°. Portevin and Durand[1] indeed found this angle very nearly 138°, by optical measurement applied not to the section but to the etched surface. This angle, 138°, has no evident crystallographic meaning, and, in view of the irregularity shown in Fig. E of Plate 36 it may only be an accidental one.

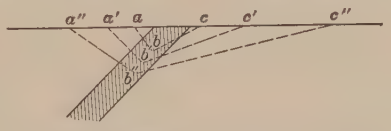

Fig. 100.—Suggested course of etching of the Neumann bands.

None of the five directions which the intersections of the 211 planes can make on such a section, shown by the broken lines 1 to 5 in Fig. F of Plate 36, corresponds to the direction of the sides of these furrows, 1 being flatter and 2 to 5 steeper than they. This is direct evidence, because this micrograph is of the very section in which these etching furrows of Fig. E of Plate 35 occur. The lack of correspondence between the sides of these furrows and the crystallographic directions suggests that the etching takes place by steps like those sketched at *abc*, *a′b′c′*, and *a″b″c″* in Fig. 100. This inference is strengthened by Stead's observation that a groove cut mechanically on a cube face widens out in this way on etching.[2]

Additional evidence of the irregularity of these etching grooves is given by their appearance in cross-section on the right-hand side of Figs. F and G of Plate 35. These show two faces of a rectangular prism of thermit steel etched very deeply after Neumann bands had been developed in it abundantly. Hence these right-hand sides show the cross-section of the etching grooves on a face normal to the plane of the paper.

578. A ladder arrangement such as is shown in Fig. B of Plate 33 often occurs, with rungs sometimes running as in this case between parallel stringers, and sometimes between stringers which are inclined to each other. The rungs in this case may be slip bands. This tendency to stop on reaching a Neumann lamella of another system is shown also in Fig. F of Plate 34. This may be an etching effect, for the Neumann lamellæ

[1] *Rev. de Metallurgie, Mem.*, 1914, vol. 11, p. 775.
[2] Private communication, Aug. 30, 1915.

on unetched polished surfaces show this tendency but slightly if at all. But even as an etching effect it is hard to explain.

579. Their profiles, as disclosed by copper plating by Rosenhain's method the surface of a specimen in which they have been developed, and then cutting, polishing, and etching a section across it, are steps such as are shown in Figs. A to D of Plate 36, recalling the sharpest of the steps of the supposed slip bands of Fig. E of Plate 25.

The behavior of Neumann bands under oblique illumination tallies with this step shape of their profiles. For when a given Neumann band has, by rotating the stage of the microscope, been brought to a position in which it shines, I find that it does not reappear on 180° rotation about the axis of the microscope, as it would in case it were a rounded furrow or ridge in profile, and as an occasional one would if it were a plane-sided ridge or furrow (§469). So, too, if a polished surface on which Neumann bands have formed is next heated till it turns blue, and is then polished very lightly, I find that light bands apparently coinciding with the Neumann bands appear on rotation to a certain position under oblique illumination, but do not reappear on 180° rotation. Such bands would naturally be caused by rubbing off the blue at the "high lights" or salient angles.

Each of the Figs. C and D of Plate 35, besides being a plan of one face of the rectangular prism, shows a profile of one or more of the other faces.

580. Lack of Sharpness in these Profiles.—Though on moderate magnification the lightly etched profiles look very regular, as in Fig. A of Plate 36, thus agreeing with the evident crystalline nature of the deformation by which the Neumann lamellæ are formed, on greater magnification, as in Fig. B and especially in Fig. C they are seen to be much rounded. Further study is needed to show whether this represents only unintentional flowing of the surface in polishing these sections, or whether it is due to some other cause, such as fluid motion, or a succession of slips along planes parallel to the Neumann lamella and integrating in profile into these curves.

581. Suggestions Given by Etching as to the Relation between the Profile Steps and the Direction of the Neumann Lamellæ, Figs. A to D of Plate 36.—When Neumann bands are developed on a polished surface, and are then shown in profile by cutting sections across them such as those of Figs. A to D of Plate 36, after copper plating the initial polished surface so as to preserve these profiles, they are invisible on the cross-section. But because each band is only the outcrop of a stratum or lamella as we shall shortly see, this lamella ought to be developed on this cross-section by etching it, and each lamella should then correspond to one of the risers of the steps in the profile. The suggestions of such lamellæ, seen at *M* and *N* in Figs. B, C, and D of Plate 36, are probably misleading, because they match the steps in the profiles poorly. That in Fig. B is narrower than the riser, and moreover lacks the straightness of Neumann bands. The matching in Fig. C is bad. Moreover, whereas in Fig. B and C the apparent

lamellæ point to the right, *M* of Fig. D points to the left. It is technically difficult to show the relation between the profiles of the Neuman lamellæ and their path across the section, because etching long enough to show this path clearly etches the contact between the copper and the steel so deeply that the profiles themselves cease to be detectable.

582. Fringes, which may be interpreted provisionally as a special form of Neumann lamella, are shown very strikingly in Fig. B of Plate 45,[1] by Stead, as dark streamers running to the left just above *a*, and downward just below it. Other streamers which are black near *e* and run thence upward and to the left toward *c*, seem to pass taperingly into an invisible stage, and to reappear as white near *c*. A like change from light to dark through invisibility is shown by the nearly horizontal streamers in Fig. A of Plate 45, and from light through dark to invisibility in Fig. AA of Plate 30. This last is evidently a twin, probably congenital.

The streamers in Fig. A of Plate 45 also seem to be twins. In view of the aurora-like figuring of these streamers in Figs. A and B of Plate 45, we next interpret as like streamers the rays near *d* running about N-60°-E in Fig. B, and also the light and dark ribs running out of the great white spinal column *b*.

Osmond[2] shows like fringes on the polished and etched surface of a crystal of steel of 0.05 per cent. of carbon, and I find like ones in the twisted crystals of native copper of which parts are shown in Figs. D, E, and F of Plate 45.

That these streamers are twins is suggested strongly by the sharp contrast in brightness between them and the ground mass, and more particularly by the fact that the whole width of a given streamer is of uniform brightness at any one part of its length, and that of any given intervening streak of ground mass is of another but also uniform brightness, with a perfectly abrupt change of brightness at the edge of the streamer. Further, Osmond's streamers not only turned into Neumann bands, but also had etching figures which showed strong signs of twinning.

The fringes in the copper crystal were not present before the twisting, which therefore both created and curved them. Thus if twins they are evidently mechanical. The same is probably true of these other cases.

[1] Figs. A and B of Plate 45 are sections of a bar cut from a meteorite, twisted about its longitudinal axis, and then cut in two longitudinally, polished, and etched. Dr. Stead generously permits me to make this first publication of them.

[2] Osmond, Fremont, and Cartaud, *Rev. de Metallurgie, Mem.*, 1904, Vol. I, Fig. 18, Plate I, opposite p. 24, and Figs. 17, 20, and 22. They recognized that these streamers turned into Neumann bands, but seemed unwilling to assert that they were twins, pointing out that "geometrical regularity is lacking and, whatever may be the internal mechanism of the deformation, it has not to do with twins in the classical sense of the word" (p. 24). They identify the fringes with figures found in an alloy of gold and rhodium by Roberts-Austen and Osmond. See *Phil. Trans. Royal Soc.*, 1897, A, vol. 187, Plate 10, Fig. 12; and Heyn, "Ueber das Kleingefüge des schmiedbaren Eisens, besonders Stahles," *Stahl und Eisen*, 1887, vol. 8, p. 82, and "Umwandlung des Kleingefüges bei Eisen und Kupfer," *Zeit. Verein deutsch. Ing.*, 1900, vol. 44.

583. Thorns and boundary edgings[1] are unexplained deep-seated markings developed occasionally in low-carbon steel, but only by etching and only after plastic deformation, and therefore clearly representing neither mechanical defects nor local concentration of impurities.

Fig. C of Plate 45 shows both these markings. Of it we are told that spikes sharp as needles, or widened at their bases like rose thorns, often start from the grain boundaries and from their edgings. From these words and from other micrographs of this memoir I infer that the isolated markings *SS* and those at *R* are thorns, and from these that the markings at *Q* also are. It is true that the words "slashed appendices festooning the grain boundaries irregularly" by which the "edgings," are described, fit the markings at *Q* better than those at *T* which I take to be the edgings; yet a comparison of the latter with the edgings in other micrographs of this memoir confirms me in interpreting *Q* as thorns.

Osmond connects the thorns with the crystalline structure of the steel, because he finds them oriented parallel to the faces of a cube, but the edgings he connects with the network structure, first because it is the network itself that they border, and second because, in case there are both coarse and fine grains in a given steel, the edgings are thick in the small grains and thin in the large ones. Without further elaboration this reasoning is not convincing. Grain boundaries are very often bordered with ferrite or cementite offshoots which evidently follow the crystalline structure of the grain in which they lie.[2] Hence attachment to the grain boundaries is wholly compatible with being connected with the crystalline structure of the grain enclosed by the boundary. And, as regards the coarse and fine grains, we should remember that in any given section the grains which look small may well be only those which are cut near one of their ends by the section, and the large ones may be those which are cut nearer their middle.

I have met what I take to be both edgings and thorns in deformed low-carbon steel, but they certainly are rare. Even Robin, who had unusual advantages for following up Osmond's work, reports that they are rather difficult to identify and even to find.[3]

[1] "Epines" and "bordures," called also "spikes" and "bordered boundaries," Osmond, Fremont, and Cartaud, *Rev. de Metallurgie, Mem.*, 1904, Vol. 1, p. 11. See in particular pp. 28 *et seq.*

[2] The orientation of these offshoots may be Widmanstättian and thus related to that of the grain in which they occur, as in the upper part of the central grain of Fig. F of Plate 9, or it may be related to that of the network as shown by its being the same on both sides of that network, for instance of that running about N–15°–W, ¾ in. above the bottom of Fig. H of that plate. Hence Osmond's edgings may or may not be related to the network which they border. (See Howe and Levy, *Proc. VI Congress, Internat. Assoc. Testing Materials*, N. Y., 1912, II, 4, p. 12 and Fig. 4).

[3] "Traité de metallographie," Hermann et Fils, Paris, 1912, p. 183. Elsewhere (*Rev. de Metallurgie, Mem.*, 1911, vol. 8, p. 441) he reports finding edgings, spikes, and light foldings, "exceptionally" on the bases of cylinders of low-carbon steel (C, 0.07) crushed apparently by shock at 475°. He does not find them in crystalline phosphoric ferrite, but suggests (*idem*, p. 447) that forging causes them in that which is not phosphoric, as if they were due to deformation of the grains along their boundaries, which is prevented in phosphoric ferrite by the weakness of the intergranular cement.

The shape of the thorns, recalling that of the deformation figures in sorbite, Fig. D of Plate 27 and through this a conchoidal fracture, suggests rather fluid than crystalline movement. Yet the details of which this shape is an integration may well be crystalline.

Rosenhain[1] refers the edgings to the existence of mixed orientation along the grain boundaries, leading to irregularity of yielding here under deformation. But this theory must be accepted cautiously. It it were true, then a corresponding deformation of the slip bands would be expected in this same area. But it has not yet been shown that there is an area as wide as these edgings, in which the slip bands show any corresponding irregularity (§426).

We really know very little about these markings. Indeed the thorns may prove to be only a variant of the Neumann bands, with irregular instead of parallel sides. For instance, the shape of the fragments of Neumann bands of Fig. F of Plate 36 is much like that of the isolated thorns at *S* in Fig. C of Plate 45. This interpretation of the thorns tallies with the inferences which I draw from Osmond's work, that their creation, like that of the Neumann lamellæ, is favored by coarseness of crystallization and by shock. The apparent difference that a blue heat favors the creation of thorns but opposes that of Neumann lamellæ remains to be studied.[2] A possible explanation is that a blue heat tends to give these lamellæ the thorn form.

584. Conditions under Which Neumann Lamellæ Form.—They are caused by plastic deformation, and the more readily the more brittle the iron, whether this brittleness is inherent in the metal itself, as for instance that caused by coarsening, by the presence of phosphorus, etc., or is caused by the conditions of the deformation, such as its suddenness or its occurring at low temperature. At room temperature, for instance, though they have in general been developed in ferrite only by shock,[3] I found that they could be developed even by quiescent pressure in the extremely brittle, very coarse, 4 per cent. silicon steel.

Again, Osmond and Cartaud[3] found them in Hadfield's Swedish iron broken by quiescent tensile rupture at the temperature of liquid air, −185°, but at room temperature they could not cause them without shock, nor could they cause them at all at a blue heat or at any higher temperature. In my A.R.M steel they form readily on sledging at room temperature. Robin was unable to form them by moderate shock in wrought iron or very ductile steels except at −185°, yet he found them in steel of 0.07 per cent. of carbon which had been struck, apparently when at room temperature, by a projectile moving at the rate of 1,246 ft. (380 m.) per second.[4] He records lines suggesting Neumann bands formed at 100°. In extremely brittle steel

[1] *Journ. Iron and Steel Inst.*, 1904, No. I, vol. 65, pp. 358 to 365.

[2] Osmond, *op. cit.;* Osmond and Cartaud, *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, pp. 467–468.

[3] *Idem.*

[4] *Rev. de Metallurgie, Mem.*, 1911, vol. 8, p. 443.

with 1 per cent. of phosphorus he formed them by shock at 20°, 300°, 500°, and even at 600°, but not at 700°.[2]

So, too, Stanton and Bairstow[2] found that breaking ductile Swedish iron with a single severe blow, or with a small number of moderate blows, caused Neumann bands which were made visible by etching, but breaking it with a great number of light blows did not.

In all these cases the readiness with which the Neumann lamellæ form is inversely as the degree of deformation which occurs. The readiness with which they form abundantly in brittle metal is illustrated by Fig. D of Plate 41, of a specimen of cast bismuth, apparently deformed only in initially breaking it, and later by spreading a crack started by means of a sharp sawcut.

Their presence in meteoric iron may be referred rather to one or more of the many collisions with other meteors encountered in the secular voyage through the solar system, or to the final impact on reaching the earth, than to the cooling strains, because the rate of cooling is so immeasurably slow, and the thermal gradient of the mass hence so flat, that important cooling strains are hardly to be looked for. The impact which causes them cannot be so violent as to raise the temperature greatly, because that would efface if it did not prevent them.

Coarsening the silicon steel with which I experimented increased the regularity of the Neumann bands.

585. The removal of the Neumann lamellæ occurs rapidly on heating meteoric iron to 800°,[3] and more slowly at lower temperatures,[4] so that after such heating they no longer reappear on repolishing and re-etching. I find that those of artificial iron do not thus disappear at 800°, but that they do on an exposure of 2 hours, and perhaps less, to 950°.

This may mean that transforming the ferrite into gamma iron or austenite effaces them, for the large nickel content of most meteoric iron would lower the temperature, *Ac*3, at which this transformation occurs to below 800°.

This effacement cannot be due to a boundary migration of the usual type, such as is thought to accompany if not indeed to cause the common form of grain growth, because these meteoric irons in which it has been observed habitually consist of only a single grain. It might be referred to the breaking down of the initial crystalline structure, deformed at the time of the formation of these lamellæ, and the growth of new grains out of the ruins

[1] *Idem*, pp. 445–448. In one passage this temperature is given as 660°, but in two others as 600°. I believe 660° to be a misprint. The intergranular brittleness of this very phosphoric steel reached a minimum at about 300° and a maximum at about 500°, and decreased rapidly with further rise of temperature, so that the steel could be forged apparently readily unless much slag was present.

[2] "The Resistance of Materials to Impact," Stanton and Bairstow, *Engineering*, 1908, vol. 86, p. 731; *Proc. Inst. Mech. Engineers*, Nov. 20, 1908.

[3] Berwerth and Tammann, *Zeits. anorg. Chem.*, 1912, vol. 75, p. 145; Tammann, "Lehrbuch der Metallographie," Voss, Leipzig, 1914, p. 129.

[4] Stead, *Trans. Faraday Soc.*, 1915, vol. 10, p. 275, and private communication, Nov. 20, 1914.

of the old, as in the common reheating of plastically deformed metals,[1] were it not that we do not know that this breakdown occurs after such very slight deformation as seems to accompany the formation of these lamellæ.

My suggestion, that this effacement results from the ferrite-austenite transformation, agrees with the relative immunity of twins in other metals toward heating (§568), for these metals in general undergo no transformation on heating at all comparable in degree with that of iron.

Occasional effacement of these bands might be referred to grain growth or to their being not true twins but bands in which the change of orientation has not been to a symmetrical or twinned position.

586. The Neumann Bands are Outcrops of Strata.—Osmond and Cartaud[2] showed that they represent continuous strata by tracing individual bands on two more faces of a crystal, by the device of mounting on a model the photographs which they had taken of the six polished and etched faces cut from a single crystal of ferrite. Stead[3] confirmed this by showing that these bands are continuous on all sides of a cylinder cut from a meteorite, as sketched in Fig. 101. Equally conclusive, if less striking, is their continuity across the section of his twisted meteorite crystal shown in Fig. A of Plate 45.

Fig. 101.—Neumann bands revealed as continuous strata by etching a cylinder turned from a meteorite. Stead.

587. Belief that They are Superficial.—The foregoing evidence disproves the belief[4] that Neumann bands are superficial and represent twins formed by the action of the emery used in polishing, when its path happens to coincide with one of the planes of cleavage of the specimen. It is hardly conceivable that the path of the emery should circle round the cylinder in the way sketched in Fig. 101, or that it should curve in the way shown in Fig. A of Plate 45. But complete disproof of this belief is given by the appearance of the Neumann bands in Figs. B, C and D of Plate 35, on the wholly unpolished cleavages.

For that matter, the theory that they are superficial is inconsistent with the fact that etching does not show them except after the specimen has been deformed. In silicon steel, indeed, etching may develop them even when there has been no intentional deformation, but this is to be referred to the ease with which even unintentional deformation causes the Neumann lamellæ themselves.

[1] See Stead, *Journ. Iron and Steel Inst.*, 1898, No. I, vol. 53, p. 145, who found that grain growth of deformed iron occurred at 500°, and more rapidly between 600° and 750°; Heyn, Martens-Heyn, "Materialenkunde," II, A, 1912, pp. 229 to 233, who found that this breakdown occurs between 417° and 616°; Sauveur, *Proc. VI Congress Internat. Assoc. Testing Materials*, N. Y., 1912, 1st Section, II, 6, who found rapid grain growth at 650°; and Chappell, *Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, p. 465, who found that disintegration of the grains began at a temperature which was the lower the greater the deformation, and sometimes as low as 350°. I have noticed an apparently complete disintegration of the crystalline structure on reheating deformed metal to 600°, but never on reheating to 550°. (See note to §492.)

[2] *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, p. 466.

[3] Private communication, Feb. 16, 1915.

[4] Le Chatelier, *Rev. de Metallurgie, Mem.*, 1914, vol. 11, p. 766, cited by Matweieff.

588. Neumann bands are distinguished from slip bands, which often accompany them on polished surfaces of ferrite, by being of a higher order of width and by their stiff look, which contrasts strongly with the usual waviness of the slip bands. These are rather like rods of boiled vermicelli, parallel sided but sinuous, whereas the Neumann bands are blade-like. For instance, in Fig. E of Plate 32 the hairy lines between *b* and *c* are slip bands, the straight lines in grain *abd* and probably the slightly curved ones in area *dbe* are Neumann bands. In Figure D of Plate 33 the ripple marks are slip bands, the stiff areas *MM'* and *NN'* are Neumann lines.[1]

A further and very important difference is that, whereas the slip bands, especially if caused by moderate deformation and after heating to say 100°, very often fail to reappear on repolishing and etching, the Neumann bands reappear persistently, and apparently are not effaced except by heating to above the transformation range, that is, above *Ac*3 (§585).

Because of the difference in their order of magnitude, Neumann bands on small magnification, such as *a*, *b*, *c*, and *d* of Fig. D of Plate 31 and those in E of Plate 32, often look as the slip bands do on larger magnification, as for instance slip bands *a*, *b*, *c*, *d*, and *e* in Fig. E of Plate 33.

It is too early to say that any or all of these tests can in every case distinguish slip bands from Neumann lines. It is still very possible that some of the lines which have been taken for Neumann bands are slip bands and *vice versa*. The formation of the Neumann bands may indeed be accompanied by slip along their plane (§612).

589. To what Crystallographic Faces are the Neumann Lamellæ Parallel?—In ferrite Linck[2] found the angles 114° 6′ and 144° 44′ between the surfaces of the Neumann strata and the faces of the containing cube, and 131° 49′ between the surfaces of the Neumann strata themselves. These angles imply that these strata are parallel to the faces of the crystal called the {211} or leucite trapezohedron, Fig. 37 (§357). Osmond and Cartaud[3] confirmed this, and in the next section I give evidence, some of which indicates that the cubic cleavages also may be followed.

590. Crystallographic direction of the Neumann bands, the outcrops of the lamellæ. The lamellæ, in following the planes of the {211} trapezohedron, would intersect a cubic face either parallel to one of its diagonals,

[1] In Fig. D of Plate 31 certain of the gray areas are seen in the photographs to consist of very fine hair lines, imperfectly reproduced by the half-tone process and contrasting very strongly with the stiff straight Neumann lines which pierce the letters *a*, *b*, and *c*. In Figure C of Plate 33 the broad bands *M* and *N* are Neumann bands, the wavy lines are slip bands.

[2] Linck, *Zeits. f. Krystallographie u. Mineralogie*, 1892, vol. 20, p. 212. Tschermak in 1874 found angles of approximately 109° 28′ and 131° 49′. These angles could be explained by assuming that these strata are parallel to the faces either of the crystal called the tris-octahedron {221} or to the faces of a cube twinned with the cube of the iron itself in the way in which interpenetrating cubes of fluorite are twinned, but for other reasons he rejected the former of these explanations (*Berichte Wien. Akad.*, 1874, vol. 70, p. 443). In the 6th edition of his "Lehrbuch der Mineralogie," 1905, he mentions this latter interpretation but not the former. Yet he appears to accept Linck's interpretation given above. Linck's angles are inconsistent with both of Tschermak's interpretations of 1874.

[3] *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, pp. 466, 467.

as on the top of the cube of Fig. 119 (§838), or parallel to a line running from one corner to the middle of the opposite edge, and thus making angles of 26° 34′ and 63° 26′ with the two edges, as on the front and right-hand face of that cube. This can give rise to six[1] directions only on any cubic face. The intersections of these 211 planes follow nine directions on an octahedral face and five on a rhombo-dodecahedral one, as shown in Figs. 47 and 50 (§447, 448). Evidence that the Neumann bands follow these directions on these crystallographic planes is thus evidence that the lamellæ are parallel to the 211 planes. Such evidence is given by the micrographs of Plates 34, 35, and 36. Let us glance at it.

In Fig. A of Plate 34, which shows the Neumann bands on what is approximately a cubic face, the silhouette or crude round-armed cross must needs lie with its arms running approximately diagonally of the square of which a perfect cube face would consist, as found by Osmond and Cartaud (Fig. E of Plate 26). Having thus established the directions of the sides of the square, we next find that the various Neumann bands, lettered here from *A* to *I*, follow all six of the directions which the intersections of the 211 planes make on such a face, and in addition one, *G*, parallel to two sides of the cube face, and two others, *F* and *I*. To these last two let us return later.

This number of directions, nine, is unusually great. More than four directions are rarely found in a single field. This large number of directions I refer to the ease with which Neumann lamellæ form in this extremely crystalline material. Yet it is possible that the etching discloses here not really developed Neumann lamellæ, but only partings such as those lamellæ follow when they form, and hence lying along the 211 and cubic planes. For though etching has not been shown to develop what look like Neumann bands in other forms of ferrite except after deformation, and though even in this silicon steel there are certainly more Neumann bands after than before deformation, yet there is the possibility that that material even without known deformation may yield etching bands along the 211 planes which thus simulate Neumann bands.

That these etching bands follow these crystallographic directions is shown by reproducing them in Fig. C of Plate 34, not with their theoretical geometrical relation to the square, but parallel to their actual directions in the micrograph, and by lettering them in the same way as in Fig. A of Plate 34.

Considering the deformation which the surface has undergone in the development of these numerous lamellæ, and that this may not be accurately a cube face, this agreement with the theoretical directions is perhaps as close as can be expected. In a second field in this same surface the agreement is even closer.

Figs. B, C, and D of Plate 35 show the position of the Neumann bands developed by deformation on the cubic cleavage faces of a prism of thermit

[1] Six, not ten. The directions of the lines running from any given corner to the middle of the opposite face are parallel to those running from the corner diagonally opposite.

steel. In Figs. C and D the cube edge is seen plainly running across the top of the figure.[1] These two figures taken together show all six of the 211 directions and no others, save one, shown at *G* in Fig. B on larger magnification, parallel to two sides of a cube face.

Further confirmation is given by the direction of the grooves formed along the Neumann bands by etching the three cubic faces of a rectangular prism cleft out of a mass of thermit steel, shown in Figs. E, F, and G of Plate 35. In each of these the white broken lines show one or more edges of the prism. These grooves here follow all six of the 211 directions and no others, save that Fig. E has a wide white vertical cubic band reaching down nearly an inch from its upper edge.

Still further confirmation is given by the Neumann bands of Fig. F of Plate 36, developed by etching a rhombo-dodecahedral truncation of the prism just referred to. They follow all five of the possible directions which the intersections of the 211 planes on such a section can make (§448), as is shown by their parallelism to those intersections drawn here in broken lines alongside the rather vague dark Neumann bands.

591. Cubic Neumann Bands.—The two cases of cubic Neumann bands just mentioned conflict with the belief hitherto held generally that the directions of the Neumann bands are only those of the {211} trapezohedron. These two cases seem to me clearly cubic, and so does band *G* of Fig. A of Plate 34. That is to say, the intersections of the cubic planes on a cube face must needs be parallel to the square sides of that face, but neither those of 211 nor of octahedral planes can be (§§836, 838 and Figs. 43, 44). In Fig. A of Plate 34, band *G* is thus parallel to the top and bottom of the cube face implied by the direction of the silhouette or cross and by those of the other Neumann bands. The appearance of *G* and its behavior on etching are as much like that of the others as could be expected in view of the different inclinations of their planes to that of the microsection. Hence if the six bands following directions *A* to *F* and *H* are Neumann bands, which seems certain, band *G* also probably is. In the same way band *G* in Fig. B of Plate 35, which is parallel to two sides of a cube face, is much like the undoubted Neumann band *F* running diagonally of the same face.[2]

It is not inherently improbable that Neumann bands, though habitually parallel to the planes of the 211 trapezohedron, should thus occasionally be parallel to cubic ones. Hence I interpret these provisionally as cubic Neumann bands.

592. Maximum Number of Directions of Neumann Bands.—Because these 211 planes follow 12 different directions and the cubic planes three, none of which are parallel to any of the 211 planes, it follows that on random sections the Neumann bands might conceivably follow as many as 15

[1] In Fig. D of Plate 35 the cube edge is shown artificially by means of the broken white line.

[2] I have to thank my colleague, Professor Alfred J. Moses, for bringing my attention to the cubic direction of the Neumann bands in this specimen, as well as for much valuable advice and information connected with this study.

directions. Moreover, adding the cubic to the 211 planes increases the possible number of directions on a cubic face to eight and that on a rhombododecahedral face to six, but does not increase that on an octahedral face (Figs. 46, 47, 48, and 50, p. 271).

The eight directions which Matweieff[1] shows for supposed Neumann bands are open to doubt, first because certain of his supposed Neumann bands fail to change direction on passing beyond a grain boundary into another grain, and second because they disappear on deep etching. On both these accounts they should be interpreted rather as polishing scratches, as Portevin and Durand[2] reasonably point out. But neither of these objections applies to the nine directions shown in Fig. A of Plate 34.

593. Curvature and Variance of Neumann Bands.—Bands *B* and *I* in Fig. A of Plate 34 remain to be accounted for. *B* is very nearly parallel to *F*, making with it an angle of about 4°. Moreover *B′*, which might be taken to be parallel to either *B* or *F*, curves so that its two ends make an only slightly smaller angle, 2°, with each other. In another field there is a like line making an angle which varies from 2° to 4° with the bands corresponding to *B*. The curvature is not greater than that found in nonmetallic crystals. *B*, therefore, may be interpreted as a variant of *F*. The curvature may perhaps be due to deformation of the specimen after *B* formed, a deformation which need not affect the other bands, either because prior to them or because of its direction. The variance may represent the deviation of this section from being a true cubic face (§592). By like reasoning *I* may be taken as a variant of *E*.

Or, reversing our point of view, we may next take *B* and *E* as the diagonals of the cube face, as sketched in Fig. C, and *F* and *I* as variants of them, and then sketch in the rest of the lines. Having done this, we find that the arms of the cross are not strictly parallel to the diagonals *B* and *E*, but the error is hardly too great to be explained by the crudity of the cross and the possible deviation of the section from being a true cubic plane.

594. Relative Frequency of the Diagonal and 2 : 1 Directions.—Because each lamella cuts two cubic faces at 26° 34′–63° 26′ and only one diagonally, the former direction ought to be twice as frequent as the diagonal, taking the three cube faces collectively, provided that the Neumann bands in these two directions are equally prominent. Of course on any one face this need not hold true, for the stresses might even result in creating Neumann strata in one direction only, with intersections all diagonal on one cube face and all parallel to the directions *tl* and *tv* on the other faces of Fig. 119 (§839, Appendix I).

595. Neumann Bands and the Widmanstättian Figuring.—The Widmanstättian figuring follows the octahedral planes, the intersections of which are diagonals of a cubic face and thus coincide with two of the six 211 directions of the Neumann lamellæ on such a face (Figs. 43 and 44).

[1] *Rev. de Metallurgie, Mem.*, 1914, vol. 11, p. 767, Figs. 1 and 1*b*.

[2] *Idem*, p. 774

The octahedral intersections coincide also with three of the nine 211 intersections of an octahedral face (Figs. 46 and 47) and with one of the 211 intersections of a rhombo-dodecahedral face (Figs. 49 and 50). Hence the Neumann bands ought to be parallel to the Widmanstättian figuring in a considerable proportion of cases.[1]

596. Parallelism of the Neumann Bands to the Slip Bands.—The belief that both the slip planes and the Neumann lamellæ usually follow the 211 planes (§449) may suggest that their traces, the slip bands and the Neumann bands, must be parallel. Yet this relation does not in itself imply parallelism, because there may be no less than 12 of these 211 directions on a random face, of which some may be followed by the Neumann lamellæ and others by the slip bands actually developed, especially if, as is usual, the two are developed by deformation at two distinct periods.

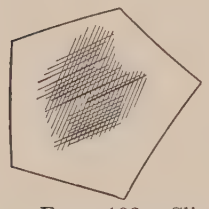

FIG. 102.—Slip bands run parallel to Neumann bands. Osmond and Cartaud.[3]

This parallelism in ferrite is illustrated in Fig. D of Plate 31. Note that above the Neumann band *a* is a series of extremely fine lines parallel to it. These are probably slip bands. Again, there is a band about ½ in. wide shortly below *d*, covered with fine hair lines running parallel to the two broad black N-60°-W lines which *d* crosses. The fine lines are probably slip bands and the broad ones Neumann bands.

Osmond and Cartaud[2] found that this parallelism occurs also in chrome-nickel austenite, and it probably occurs in austenite in general. Thus they noticed that if slip bands are developed in a field in which Neumann bands are already present, one set of the former is parallel to the latter, as shown in Fig. 102, in which the more flatly inclined set of the slip bands, shown by narrow lines, is parallel to the Neumann lines, here drawn as broad lines. Because the slip bands in austenite are octahedral, we infer from this parallelism that the Neumann bands also are.

597. Direction of the Neumann Lamellæ Relative to the Stress which has Caused Them.—Robin[4] found them normal to the direction of impact of a projectile, whereas when they are formed by the blows of a sledge in my silicon steel they run by preference at approximately 45° with the direction of impact. This is to be interpreted as meaning that, of their possible directions, they prefer that most nearly at 45° with the greatest shearing stress (§§401 *et seq.*). On the base of a cylinder hammered on its other end Robin found two sets of Neumann bands near the center, but near the circumference only one, apparently in any given grain.

[1] Matweieff (*Rev. de Metallurgie, Mem.*, 1914, vol. 11, p. 770) holds that they are parallel to the Widmanstättian figuring without qualification, but in this I believe that he must be in error.

[2] *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, p. 468.

[3] The narrow lines represent slip bands, the broad ones Neumann bands or narrow twins, *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, Plate XLVIII, Fig. 22, opposite p. 448.

[4] *Rev. de Metallurgie, Mem.*, 1911, vol. 8, p. 444.

598. Parallel Neumann Lamellæ have Identical Internal Orientation.— By their "internal orientation" I mean the orientation of the constituent particles of any given Neumann lamella, as distinguished from the orientation of the lamella itself as represented by the inclination of its two main surfaces.

We have seen that each lamella is usually parallel to the faces of the {211} trapezohedron, but this tells us nothing of the orientation of the crystalline particles which compose it. We may assume provisionally that this orientation is constant throughout any one lamella, and our present question is whether it is also the same in the various members of a given parallel series.

To fix our ideas we may call the inclination which the surface of the cylinder of Fig. 101 bears at any point to this internal orientation the "surface orientation" at that point. Clearly this must vary continuously along the ring which the outcrop of any given lamella forms on the surface of the cylinder.

For simplicity of presentation let us assume that, at one point in this ring, the surface of the cylinder is parallel to the octahedral planes of that ring, and at another point is parallel to its cubic cleavages. If, still for the purpose of fixing our ideas, we assume that Stead's observation that the octahedral planes are attacked more readily than the cubic applies to this material, then the former point will etch deeper than the latter.

Just as the octahedral planes differ from the cubic in their solution pressure, so the solution pressure must vary continuously along each ring to match the continuous variation in the surface orientation. What is true of these rings is also true of the rest of the metal, the matrix in which these Neumann lamellæ lie.

At the nodes or gaps in these rings, shown in Fig. 101, the Neumann lamellæ become very faint, as they do between their bright parts at the left of Fig. A of Plate 45 and their dark parts at its right. The nodes shown in Fig. 101 lie in a straight line parallel to the axis of the crystal. At 180° from this on the other side of the cylinder runs a like line of nodes.

This faintness or indeed invisibility means that the tint and brightness of the ring here are identical with those of the enclosing metal or "matrix," because visibility means only that the object seen differs in tint or brightness from the surrounding ones, and invisibility means only that it does not. Hence at a node the ring is identical in tint and brightness with the matrix, and, because both have been discolored by the etching reagent, this means that they have etched alike, and this means that there is some specific relation between the surface orientation of the ring and that of the matrix which gives identical etching tint and brilliancy.

Because the cylinder was cut with its axis parallel to that of the crystal, the matrix lying in the line of the nodes, or in any other line drawn on the surface of the cylinder parallel to that axis, must have uniform surface orientation along its length. Because this line of the nodes is thus parallel to the axis of the cylinder, the surface orientation of all the Neumann strata

must at this point bear to the constant surface orientation of the matrix along this line, that specific relation which gives invisibility. Hence all these lamellæ must here have identical surface orientation at the line of the nodes and hence, too, they must have identical internal orientation, which is what we started out to show.

Having proved this we can, indeed, go a step beyond, and hold that this specific relation which gives invisibility implies crystallographic equivalence, because only crystallographically equivalent surfaces can etch alike, or indeed behave alike in any other way. In short, that along the node line the surface orientation of the Neumann strata is the crystallographic equivalent of that of the matrix, which certainly suggests very strongly that the two are symmetrical, and hence that the Neumann bands are true twins.

599. Additional evidence from the parallelism of the variations in etching tint in forcibly enspiralled Neumann lamellæ is given by Fig. A of Plate 45. Here a bar was cut by Stead from a single grain of a meteor and twisted about its axis. This section was then cut along the axis of the bar which had been thus enspiralled, and was then etched.

The twisting of the bar forced each crystal unit in each Neumann lamella to rotate about the axis of the bar. If this motion had been strictly vectorial, each unit persistently facing in its initial direction, as so many compass needles would have, all would now be inclined alike to this section, and hence all would have etched to the same tint. However surprising this inference may be at first, it seems to me inevitable. But instead their tint varies progressively from left to right of the figure, being brighter than that of the enclosing metal somewhat at the left of the middle of their length, passing thence through a stage in which it is so nearly identical with that of the enclosing metal that they are invisible, becoming slightly darker than that metal a little to the right, and thence increasing progressively in darkness.

This progressive variation in tint indicates that the orientation of these units varies from end to end of any given Neumann lamella. Deferring to §678, our study of the light which this variation throws on the nature of the mechanism by which the metal accommodates itself to this twisting, for our present purpose the important point is that the variation in crossing this field is equivalent to that in passing round the cylinder of Fig. 101, and that here as there the invisible region of the various lamellæ occurs at one and the same part of their length, so that the initial parallelism of their internal orientation has persisted through this forcible deformation of that orientation. This, by the process of reasoning followed in the preceding section, shows that these various lamellæ have preserved the initial parallelism of their internal orientation through its forcible change by the twisting, a parallelism which we are trying to prove.

Less striking than the nearly horizontal Neumann lamellæ are those near *S* inclined at about 45°. In these too the variation in tint is progressive,

and so it is in the shorter and nearly vertical streamers at the right of *R*, in the twin in copper in Fig. AA of Plate 30, in many of the streamers in Fig. B of Plate 45, and in some of the "fringes" of Osmond and Roberts-Austen, which we have already interpreted provisionally as Neumann bands (§582). Other streamers in this Fig. B change tint more abruptly, but some of these changes width progressively.

The fact that such progressive variations in tint are often due to mere variations in the incidence of the light, does not detract from the force of this evidence. Even if this variation should here be referred to this cause, the important fact would remain that the invisible region of the several lamellæ occurs at the same part of their length, thus showing that this part has the same effect on this supposedly varying light, and hence that the initial parallelism of orientation has persisted.

600. Additional Evidence from the Zigzagging of the Slip Bands.—Zigzagging of the slip bands occurs not only at what are very probably Neumann lamellæ in ferrite, shown in Fig. F of Plate 30, but at what certainly are in the tin examined by Edwards.[1] In both cases the slip bands have the same direction in all the members of any one set of parallel Neumann bands.

In general the slip bands, on meeting a Neumann band, do not zigzag but simply stop. I have found a few cases in which, at Neumann bands of the order of width of those of Fig. F of Plate 30,[2] they change directions as they do in that figure. Sometimes on entering a zone evidently identical in nature with those in which they zigzag and with those in which they vanish without visible deflection, the slip bands deflect and then die out. But I have never found a case in which they cross a Neumann band on an unetched surface without change of direction, so that in this respect they show the necessary symptoms of twinning, in that they invariably change either their own direction or else the contrast between their treads and risers, for

[1] "Metallic Crystal Twinning," *Journ. Inst. Metals*, 1915, No. 2, vol. 14, Plate III, Figs. 8 and 9 of preprint.

[2] Two reasons for questioning the competence of this figure to prove that Neumann bands zigzag the slip bands suggest themselves: first, that its zones are too wide to represent Neumann bands; and second, that they may be annealing twins. But these reasons fail to bear scrutiny. However thin a Neumann lamella, its apparent width on a microsection must increase as the plane of that section approaches parallelism with the lamella, till, as in a photograph given me by Dr. Stead, the lamella occupies the whole field. Hence the unusual width of the lamellæ of this figure may mean simply that the angle between the plane of the microsection and the lamella is sharp.

Second, though the history of this thermit specimen raises only a probability that it has not been heated since deformation, and hence that these zones cannot be annealing twins, in another specimen of thermit steel I have found zones which, though they cannot be interpreted as annealing twins, yet cause comparable even if less pronounced twinning effects on the slip bands. This is a thermit riser from a weld. Its bright fracture shows that it has not been heated since it was broken from the weld. In this state it fails to show zones which either zigzag or efface the slip bands. After violent shock and repolishing but without heating it has such zones, which therefore must be mechanical and not of annealing origin. In a few of these the slip bands zigzag; in many they vanish to reappear on the other side; and in one at least they deflect on entering a zone and then die out, to reappear on the other side with their original direction. Thus they show the twinning symptoms.

the dying out of the slip bands is only the obliteration of that contrast (§558).

In etched sections I have not succeeded in finding the twinning effects of Neumann bands on the slip bands, probably because when seeking them I had not learnt that Neumann bands wide enough to favor this search can be developed in the A.R.M. and thermit steels.

The suggestion of zigzagging at *a* in Fig. A of Plate 33 is unfair, because here the apparent width of the lamella, with which this zigzagging corresponds, must be far greater than the true width, as is indicated by the narrowing to a node at the left of *a*. The suggestions of zigzagging of the slip bands and faulting of the Neumann bands at their intersections in Fig. C of Plate 33 need further study. The wide zones *M* and *N* of Fig. E and the narrower ones of Fig. D of Plate 33 indeed seem to cause no twinning effect in the bands *a*, *b*, etc. These latter are not Neumann bands, because they are caused by quiescent deformation which is incompetent to cause Neumann bands in this material, and hence they interpret themselves as slip bands. Yet they undergo no change on crossing these apparent etched Neumann bands *M* and *N*. This I cannot interpret confidently, but, in view of the great exaggeration which etching causes in the apparent width of Neumann bands, it may well be that their true width in these cases is so slight that the deflection of the slip bands on crossing the Neumann bands escapes detection; and moreover these may not be Neumann bands.

Again, a very narrow twin may fail to exert a detectable twinning effect on the slip bands. A slip band probably owes its position to a weak contact along a given slip plane. If this weak contact is crossed by a narrow Neumann, then the tendency of the Neumann to deflect the slip band is opposed by the tendency of slip to follow, on the other side of the Neumann, this same weak contact. This conflict might well lead the slip plane to cross the Neumann by means of a series of conjugate steps so minute as to be invisible, somewhat as rupture, in crossing a system composed of two massive blocks of wood which match exactly on either side of a thin veneer, may give no prominent token of the direction of the fiber of the veneer in crossing it.

601. The Internal Structure of the Neumann Lamellæ, though Differently Oriented, is like That of the Enclosing Metal.—This I infer from the phenomena of Fig. 101, and Fig. A of Plate 45. The invisibility of the Neumann bands at the nodes of Fig. 101 implies their identity with the enclosing metal in tint and brightness, and this in turn, in etched specimens, points to a crystallographic equivalence, including identity of coarseness. Their visibility everywhere except at the nodes implies difference of etching, which, because it is not be be explained by a difference in structure or in coarseness, is referred to difference in orientation. Hence we assume that at the nodes the Neumann lamellæ are equivalent in every way, including orientation, with the enclosing metal, but that elsewhere they are not equivalent in orientation. Because the orientation is uniform through-

out each lamella and also uniform throughout the enclosing metal, this contrast suggests that the equivalence of orientation at the nodes is not through identity but through symmetry of orientation. The same reasoning applies to the identity of etching tint in the corresponding parts of the nearly horizontal Neumann lamellæ of Fig. A of Plate 45.

It is true that the very deep etching of Portevin and Durand,[1] though it certainly furrows out the Neumann lamellæ, yet develops on them etching figures identical with those of the enclosing metal. These figures are cubes diagonal to the Neumann lamellæ themselves, thus suggesting that they are parallel to the edges of the cube face (§590). The simplest explanation, I think, is that suggested in §577 in connection with Fig. 100, that the sides of the etching furrow do not lie in the Neumann lamella but in the enclosing metal, so that the part of this furrow actually occupied by the Neumann lamella is so small that it is overlooked, and that the figuring which we note is not that of this lamella but that of the enclosing metal. This explanation is supported by Stead's finding the very figuring which Portevin shows, on cutting a groove mechanically in a cube face and etching.[2]

602. Are the Neumann Lamellæ True Twins?—They clearly differ in orientation from the enclosing metal; are they symmetrical in orientation with it? The common belief that they are is supported by the reasons which led us in §§561 *et seq.* to class other parallel-sided areas as twins, that they etch differently from the enclosing metal, that they affect the slip bands as annealing twins do (§558), and that parallel ones have identical internal orientation, as has just been shown in §598, and that because this parallelism exists after rotation, this arrest of the rotation at the same degree in all the layers suggests strongly a geometrical and hence a symmetrical degree of rotation, which is the essence of twinning.

The usual marked difference in appearance betweeen annealing twins and Neumann bands may well be referred to the usual much greater width of the former, for what are very probably broad Neumann bands are very closely like the annealing twins, as is seen on comparing the probable Neumann bands of Fig. F of Plate 30 with the annealing twins of Plate 29.

Beyond the influence of the inclination which the plane of the microsection bears to that of the lamella itself, the width of each of these classes of twins probably varies greatly from case to case. Note the variations in the width of the neighboring even-numbered zones in Figs. B and F of Plate 29, and in that of the odd-numbered ones in Fig. F of Plate 30. The absence of all faulting of the Neumann bands of Figs. A and F of Plate 34, in spite of the great inclination which certain of these lamellæ must bear to others, suggests that they are far narrower than those which have caused the steps seen in profile in Figs. A to D of Plate 36 and in plan at *G* in Fig. B of Plate 35, though this difference may perhaps represent a difference in the slip accompanying the formation of these lamellæ rather than in the

[1] *Rev. de Metallurgie, Mem.*, 1914, vol. 11, pp. 776–778.

[2] Private communication, Aug. 30, 1915.

width of the lamellæ themselves. So too, the greater width of the bands in Fig. D of Plate 32 than in Fig. A of Plate 34 may represent a difference either in lamella thickness or in throw. I here refer to the fact that the Neumann bands of Fig. A of Plate 34 are so narrow that they are not seen at all except in those parts of their length which have been widened by etching.

A difference which seems to disappear on looking more closely is that etched Neumann bands often differ from the enclosing metal in tint more than annealing twins do. But it is very doubtful whether careful examination would substantiate this difference. Note how slight this difference in tint is in the case of the Neumann bands of Figs. C, D, E, and F of Plate 33, and B and D of Plate 34.

There may be the very important difference that the Neumann bands represent translation of an order of magnitude higher than that of annealing twins, as is seen on comparing the important uplift represented by the profiles of the Neumann bands in Figs. A to D of Plate 36 with the apparent complete absence of uplift by annealing twins in spite of their usually much greater absolute width (§567). This suggests that the twinning of the Neumann bands is by wheeling, as in Fig. 103, but that of the annealing twins is by facing about. Yet the throw implied by the profiles of the Neumann lamellæ may be due not to their formation but to slip accompanying it, so that they may form by facing about (§611).

603. The Contacts between the Neumann Lamellæ and the Enclosing Metal are Planes of Low Cohesion.—This is shown by the accuracy with which the path of rupture in Fig. D of Plate 32 follows their directions.[1] The lower part of this fracture, from *D* to *E*, is parallel to the system *A* running N-20°-W. Its upper part consists of a series of long steps, of which the short treads are parallel to system *B* running N-55°-W, while the long risers are parallel to the rather faintly marked system *C* running N-5°-E.

604. Their Behavior on Meeting Obstacles.—At grain boundaries they stop or change direction, sometimes greatly, as in Fig. A of Plate 35 and at *A* in Fig. C of Plate 31, sometimes slightly as at *C* and *D* in the latter figure. In thus deflecting those which pass into any one grain remain parallel to each other. For instance the lamellæ *A*, *B*, *C*, *D*, and *E* of Fig. C of Plate 31 are parallel in the lower part of the figure, because they are all in the same grain, *b*. On crossing to the right of the grain boundary *A* and *B* in bending down remain parallel because they are in one grain *F*, whereas *C*, *D*, and *E*, though remaining parallel, change to a different direction because they enter a different grain, *G*.

The case shown in Fig. A of Plate 35 is interesting. Here the Neumann lamellæ running in the direction *AB* deflect about 60° to the left, to the direction *BC*, on entering the dark enclosed grain, and revert to their initial direction on re-entering their original grain at *C*.

[1] That the fracture may follow the Neumann bands was shown by Osmond, Fremont, and Cartaud, *Rev. de Metallurgie, Mem.*, 1914, Vol. 1, Fig. 25, p. 25.

But they do not change direction on crossing what may be taken for grain boundaries of the second order, as for instance in Fig. B of Plate 34. This is natural whether we regard these apparent boundaries as only partings between fragments of a single grain faulted thus by the shock, but without discontinuity, or as true grains of ferrite, but of common orientation because derived from a single grain of austenite.

A case in which what look like Neumann bands stop at a grain boundary is shown in Fig. F of Plate 31.

Whether they stop at a grain boundary or cross it, they show no signs of disturbance, unless for an extremely narrow distance. In Fig. C of Plate 31 of steel of 0.03 carbon, Fig. A of Plate 35 of zinc, and Fig. D of Plate 41 of bismuth, they continue quite to the grain boundaries or across them, without the slightest indication of disturbance, even when, as in Fig. C of Plate 31, they are seen under a magnification as great as 500 diameters. If those of Fig. F of Plate 31 are Neumann bands, they too show no sign of disturbance, unless their pinching out can be interpreted as such, and even this is confined to a very narrow region just beside the boundary.

605. On crossing each other one of them is often eclipsed. For instance, in Fig. D of Plate 34 the oblique one is eclipsed at the intersection, and in Figs. D and E of Plate 33 lamella *NN′* is eclipsed in crossing *MM′*. The continuous one is naturally interpreted as of later origin than the interrupted one. In addition they usually fault each other, as explained in §620, as they might whether they form by wheeling or by a facing about accompanied by slip (§612). Their intersection with slip bands is discussed in §600.

606. They do not Affect the Hardness.—This follows from Stead's observation that the removal of even a very great number of them, by heating to 800° or 900°, leaves the hardness of meteoric iron unchanged.[1] Again, though the Neumann bands may be so narrow and packed so closely in cast zinc that the structure is much like that of martensite, there is no evidence that their presence increases the hardness.[2]

[1] *Trans. Faraday Soc.*, 1915, vol. 10, p. 275, Remarks in General Discussion of the Hardening of Metals, Nov. 23, 1914.

The number of Neumann lamellæ removed in this case is estimated as "at least 1,000 in each eighth of an inch." Dr. Stead has kindly repeated these determinations and made others for me with the following results:

Specimen cut from a meteorite	Brinell hardness
In the initial state, when the specimen was a mass of twin laminæ..	131
After heating to 800°, when it was nearly free from them.....	131
After heating to 900°, when it was quite free from them.....	131
After heating to 1,000°, when it was quite free from them, but contained phosphides in solid solution, diffused from the rhabdite $(Fe,Ni)_3P$.....	137
Coarse crystals, ¼ to ½ in. diameter, from the bottom of an open-hearth furnace:	
Initial state, twinned.....	89
After removing the twins by heating to 1,000°.....	89

(Private communication, Sept. 7, 1915).

[2] Desch, "Note on Crystal Twinning and the Martensitic Structure," *Trans. Faraday Soc.*, 1915, vol. 10, p. 255.

607. The deformation of austenite and the martensitic structure have in common that they follow the octahedral planes. Moreover, if its deformation results in the creation of a very great number of parallel bands, each of them extremely narrow, the resultant pattern is just that of martensite. Such innumerable bands resembling twins may be caused in Hadfield's manganese steel, that is, manganiferous austenite, without annealing and directly by plastic deformation, as in the case shown in Fig. B of Plate 39. Here these deformation bands cut the micrograph up into rhomboids and nearly equilateral triangles, figures naturally formed by the intersection of octahedral planes with the plane of the paper[1] (§668).

A family resemblance exists between the deformation bands in this figure and in Fig. D of Plate 41, the martensitic markings of quenched carbon steel of Plate 10, A of Plates 2 and 11, and F of Plate 28[2] and the Widmanstättian figurings in Figs. A and C of Plate 28.

This resemblance to the characteristic markings of the martensite of carbon steel hardened by quenching cannot be set aside as accidental, first because in this quenching there is a cause which might well lead to deformations comparable with those which cause these markings in this manganese steel, and second because like markings are caused in other alloys by conditions like those which cause them in the hardening of carbon steel, and with results in a general way similar, as I will now explain.

That such deformations should arise in the quenching of carbon steel has already been pointed out in §243. Briefly, because the transformation from gamma toward alpha iron starts from nuclei and spreads, and because it is accompanied by great dilatation which sets up internal pressure tending to arrest further transformation by opposing the dilatation which must need accompany it, it results that though certain particles transform and swell, the rest do not; and this contrast between the swelling of certain particles and the constancy of volume of the rest causes internal deformation, which may well be so closely related to the interstratal movements which cause the innumerable deformation bands in the breaking of manganese steel, as to cause similar bands. This then suffices to explain why the difference in volume changes between different particles in quenching may cause deformation. The innumerability of the bands may be referred to the extreme intimacy of the mixture of the particles which transform and swell with those which do not transform and hence do not swell.

608. Martensitic Markings Formed in Other Alloys by Quenching.— Certain copper aluminum alloys undergo in cooling a transformation comparable with that which steel undergoes. They have martensitic markings

[1] Very striking micrographs showing this structure as caused in such steel by the deformation incidental to tensile rupture are shown by Osmond and Cartaud, *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, pp. 469–470, Figs. 23 to 25.

[2] Martensite in quenched steel of 0.17 per cent. of carbon which suggests the figuring of Figs. B and C of Plate 38 is shown by Carpenter and Edwards, *Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, p. 159, and Plate XVI, Fig. 11, following p. 176. Under oblique illumination there are parallel-sided areas in these grains which look white or black according to the direction of the light.

closely like those of hardened carbon steel if, like it, they are quenched from above their transformation range, but not if cooled slowly from above it, and moreover this quenching hardens them as it does carbon steel, though much less intensely.

Thus Breuil[1] found that the tensile strength and Brinell hardness of such alloys were nearly doubled by such quenching. Guillet[2] found that these martensitic needles in an alloy of 88 copper and 12 aluminum appeared when it was quenched from above the "first transformation point," which is "about 500° in cooling," and that they increased in importance with further rise of quenching temperature, till on quenching from 900° their structure became closely like that of quenched carbon steel, in particular reproducing its striations, so that one material might be mistaken for the other. Edwards and Carpenter[3] show a micrograph of such a quenched alloy (copper 86.8 per cent., aluminum 13.2 per cent.) which is very closely like that of the mixture of austenite and martensite formed in quenching steel of 1.60 per cent. of carbon, as shown in a micrograph which Osmond[4] published in 1899.

The connection of this structure with the hardening of steel has already been touched on in §243. These may well be mechanical twins, caused by the quenching stresses. But that twins as such can cause the accompanying hardness is improbable in view of the failure of the removal of the twins in iron to affect its hardness (§606). The twinning is probably accompanied by slip, which, by generating amorphous metal, may increase the hardness.

609. Mechanical Twins in Other Metals.—The narrow bands in bismuth shown in Fig. D of Plate 41 (§570) are almost certainly mechanical twins.[5] Habitually rather narrow twins are formed mechanically in tin. The attempt to explain these as annealing twins on the ground that tin is so mobile at the room temperature as to be self-annealing,[6] has been defeated by Edwards, who finds that these twins "can be observed with the naked eye the moment the metal is strained, and while it is still at the temperature of liquid air,"[7] though Rose has shown that the annealing of tin is confined to temperatures above about 10.[8] The slip bands zigzag in crossing these twins in the usual way.[9] Cast zinc too may contain twins, "so closely packed and so narrow as to simulate very closely the structure of martensite."[10] Whether the mechanical cause of these last twins is the casting strain or that caused in

[1] *Comptes rendus*, 1905, vol. 140, p. 587.

[2] *Rev. de Metallurgie, Mem.*, 1905, vol. 2, pp. 575 to 576.

[3] *Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, p. 166.

[4] *The Metallographist*, 1899, vol. 2, p. 261, and Fig. 1, opposite p. 257.

[5] See Desch, *Trans. Faraday Soc.*, vol. 10, May, 1915, p. 256, who shows such twinning in cast bismuth. There is no evidence that either his bismuth or mine had been annealed. Indeed the contrary is extremely probable.

[6] *Engineering*, 1913, vol. 96, p. 510.

[7] "Metallic Crystal Twinning," *Journ. Inst. Metals*, 1915, No. 2, to appear.

[8] *Trans. Faraday Soc.*, vol. 10, May, 1915, p. 282.

[9] Edwards, *loc. cit.*, Figs. 8 and 9 of Plate 3 of preprint.

[10] Desch, *loc. cit.*, and *Journ. Inst. Metals*, 1914, No. 2, vol. 12, Fig. 11, on Plate 4, opp. p. 52.

detaching and preparing the specimens remains to be proved. Edwards' work must be interpreted as a confirmation of the mechanical twinning found by Mügge in antimony and bismuth (§550), and by Wm. Campbell[1] in a large group of metals, notably in bismuth and zinc.

[1] William Campbell, "The Effects of Strain and of Annealing in Aluminium," Etc., *Appendix IV, Sixth Report Alloys Research Committee, Inst. Mechan. Engineers*, May–June, 1904, p. 881, and pp. 7 and 23 of reprint.

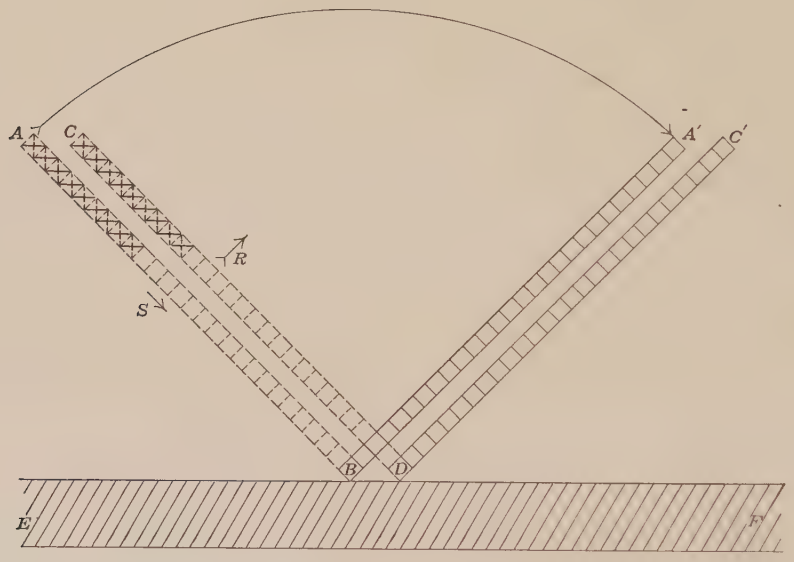

Fig. 103.—Platoons wheel to the right into a symmetrical or twinned position.

CHAPTER 24

MECHANISM OF TWINNING

610. Twinning by Wheeling and by Facing About.—Each of the soldiers of Fig. 103 could reach his twin orientation, facing southeast like *S* instead of northeast like *R*, by turning on his heel without the wheeling of the platoon on a whole, that is by simple rotation about his own axis without any translation of his body. This rotation is indicated by the two diagonals drawn across certain of the left-hand units. A diagonal drawn before rotation as one of these broken lines would, after rotation, occupy the position of the unbroken line of the same unit. This may be called "twinning by facing about" instead of "by wheeling." In §613 we will ask in which way twinning actually occurs.

611. Imaginary Mechanism of Twinning.—Let the links 1, 2, and 3 of Fig. 104 represent three crystal units extending from *b* in Fig. 94 toward *d*, in the line *bd*. The links 2′, 2″, 3′ and 3″ represent links 2 and 3 after twinning. Each link is held by means of a pin to that above it, and has on its left-hand side the ball of a Forgg[1] catch which fits into a cavity on the right-hand side of its left-hand neighbor. Similar links are supposed to stand at the right of 4, 2, and 3, and indeed to constitute the whole mass. They are omitted here for ease of exposition.

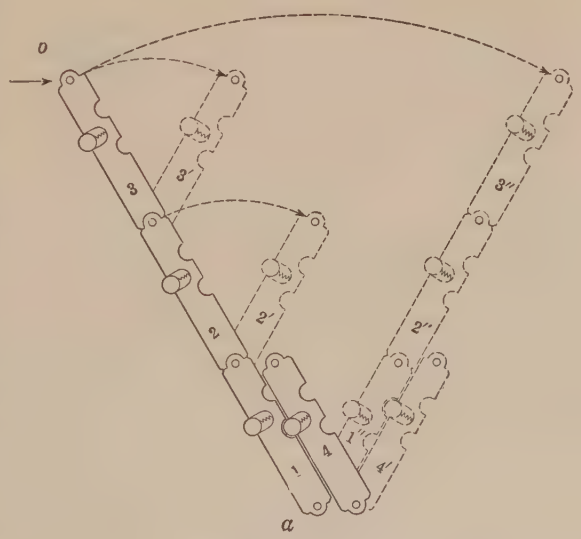

FIG. 104.—Imaginary mechanism of twinning

A knife-blade applied at *b* of this figure in the way shown in Fig. 94, p. 403, creates a component in the direction of the arrow, tending to rock these links about their pins. They cannot rock thus without forcing the several balls back into their holes against the pressure of the springs, which thus hold the whole system rigid till the stress becomes great enough to overcome them, to force the balls back, and to rotate the links toward the twinned position in which they point N.E. instead of N.W.

[1] The Forgg catches are those now used generally for the doors of small pieces of furniture. A small ball on the door fits into a corresponding cup on the jamb or sill.

We can imagine the rotation occurring either by wheeling, so that links 1, 2, and 3 wheel about the lower end of 1 into the positions 1″, 2″, and 3″, link 4 and those not shown at the right of 2 and 3 of course joining in this movement; or by facing about, so that not only link 1 but also 2 and 3 turn, each about its lower pin, into the positions 1″, 2′, and 3′, without any translation of the lower end of any link. In this latter movement the upper end of each link is supposed to be freed from its pin. Let us for the moment consider the former movement.

Once the stress suffices to start it, the rotation proceeds unopposed till, on reaching the position 1″, 2″, 3″, 4′, etc., each ball catches in the upper of the two cavities in the link next at the left, the ball of link 4′ catching in the upper hole of link 1″, and unsketched balls and unsketched links in like manner engaging the upper holes in links 2″ and 3″, and thus arresting the wheeling.

Instead of a single ball fitting into either of two cavities, we should imagine an enormous number of balls each fitting into either of two cavities, every cavity but one being filled both before and after the rotation by one or another ball, and balls alternating with cavities on each face of each link.

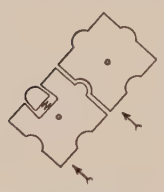

Fig. 105.—Twinning by facing about.

Facing-about twinning could occur by means of the same mechanism, but its facing about is more exact if the pivot about which each link rotates is supposed to be in its center instead of at one end, and if the links are supposed to be isometric, with their width, thickness, and length identical, as sketched in Fig. 105. Indeed the facing about the excentric pins shown in Fig. 104 might be called wheeling on an extremely small scale.

The motive force in the case of mechanical twins may be supposed to be a component of the stress which causes the deformation, and in the case of annealing twins some residual stress which has not been relieved fully by the deformation.

612. Evidence that the Neumann lamellæ form by wheeling is given by their profiles in Figs. A to D of Plate 36. These are mechanical twins at least in the case of ferrite and tin. The twinning of the corner of the calcite crystal of Fig. 94 is evidently by wheeling, for the particles initially in the lines *bd*, *ac*, and *ae* have evidently wheeled about *d*, *c*, and *e* in reaching their symmetrical positions *b′d*, *a′c*, and *a′e*, quite as the line *AB* of soldiers in Fig. 103 has wheeled about *B* in reaching its new position *A′B*. In Fig. 97 the two risers, 2 and *ja′k*, seem to have formed by like wheeling process, in which we may conceive that layers 2 and 4 have wheeled about the gliding planes which separate them from layers 1 and 3, while these odd-numbered layers retain their initial orientation. Here the system of the two treads 1 and 3, with the riser 2 between, corresponds exactly to the system of the inclined risers of Figs. A to D of Plate 36 alternating with the level treads representing the undisturbed part of the initial surface. In one case as in

the other the whole mass which is affected seems to wheel about one of its sides as a gliding plane.

Yet this is only one of several possible interpretations of the Neumann lamellæ, for each of these may instead have formed either by wheeling about a gliding plane identical with that of the riser, or by facing about, the uplift in this case being caused by simultaneous slip along that same plane. Light may be thrown on this by etching cross-sections such as those of Figs. A to D of Plate 36. This might be made to tell us the direction of the gliding plane relatively to the riser, and in addition to disclose the true width of the lamella. From these two data it should be possible to decide between the three mechanisms here suggested (§581).

A lamella of the thickness of those of Fig. D of Plate 32 would, by wheeling, suffice to cause a riser as large as those of Figs. B to D of Plate 36, allowing for the difference in magnification.

613. Evidence that annealing twins form by facing about is given by the fact that when they form they fail to disturb the previous polish of the surface (§567), which would necessarily be disturbed by wheeling. Of course we cannot here tell whether this facing about is like that of links 1, 2 and 3 of Fig. 104 in reaching positions $1''$, $2'$, and $3'$, or like that of the soldiers of Fig. 103 each of whom rotates about his own axis. In other words the difference in mechanism may be one of degree rather than one of kind; yet even so the difference is very marked.

Again if the twinning of the band $DACBE$ of Fig. E of Plate 30 occurred by wheeling to the left about B as a pivot, from the initial stratification BC into the present assumed symmetrical direction BA, how comes it that the material which has flowed in to fill this notch ABC preserves the straight line AC bounding it from the symmetrically oriented patch above? If, on the other hand, the wheeling has been to the right from A to C about B, expelling the material which previously filled this space, how comes it that the violent extrusion of this material has left the grain boundary EBC continuing past B without deflection? This is not a single case, to be set aside on the theory that here the direction of the twinning movement happens to be normal to the plane of the paper, and hence incompetent to fault the grain boundary at BC. It is a general rule incapable of being explained away thus, because the majority of sections, taken as they are at random, must be inclined to the direction of the twinning movement, and hence on such sections the twinning ought to fault the grain boundaries if it were by wheeling. But in fact annealing twinning rarely if ever does.

Like reasoning applies to the ends of twinned zones which fail to extend completely across a grain, but stop within the field of the micrographs, such as the right-hand end of zone 2 in Fig. A of Plate 29, the upper ends of zones 3, 5, and 7 of Fig. B, and the upper end of zones 1 and 2 of Fig. F of Plate 29. How can we conceive that any of these zones has reached its present position by a wheeling movement about one of its edges without causing violent disturbance of the enclosing metal at both ends? (See §458.)

If, on the other hand, the twinning were by facing about, as indicated in Fig. 105, the individual movements would be of so small an order of magnitude that they need leave no visible trace. If, for instance, each of the little squares were a crystal unit, the little disturbance caused by these particles in facing about would be no more visible under our actual magnification than the corresponding facing about of a platoon of soldiers would be from an aeroplane.

614. Twinning and Slip Band Mechanism of Osmond and Cartaud.[1]— According to these writers a twin, roughly speaking, is a re-oriented slip band, the re-orientation of mechanical twins occurring during the deformation itself and that of annealing twins during the subsequent annealing. More accurately a twin results from the re-orientation of a layer which has moved in a way which would cause a slip band on a previously polished surface, the slip band being the superficial symptom of a deep-seated movement.

This relation is seen on comparing Fig. 67, p. 338, which shows the slip mechanism of these writers, already described in §457, with Fig. 68 which shows their twinning mechanism. It is shown also in Figs. 69 and 70, described in that same place, to which the reader should here refer.

As regards the movement in the crystal unit, mechanical twinning here differs from slip first in that the rocking, instead of stopping at a random position such as $r''t''uv$, Fig. 69, reaches the position $r't'uv$ symmetrical with the initial, and hence a position of equilibrium; and second that the constituent particles of that unit themselves rotate, each about its center, into the new orientation belonging to this symmetrical position, as indicated by the narrow broken lines in the upper part of Fig. 70, instead of retaining the initial orientation, as they are supposed to during slip, and as is indicated by the unbroken narrow lines in the lower part of the figure. Thus in both respects equilibrium is reached, and after twinning the conditions are equivalent to the initial.

An annealing twin forms in two steps. First the plastic deformation, causing slip bands, gives the unit the random exterior shape $r''t''uv$ while leaving its constituent particles in their initial orientation, that of the narrow unbroken lines, Fig. 69. Then the mobility given by the high temperature of annealing both allows the unit to move accurately into the symmetrical position $r't'uv$, of Fig. 70, which it has already approached during the preceding deformation, and also allows the constituent particles of the unit to rotate about their centers into the orientation corresponding to this outward form, the orientation shown by the narrow broken lines, and thus brings the whole into equilibrium. The mobility of the annealing temperature permits the twinning, which was approached during the deformation, to consummate itself.

The application of this mechanism to annealing twins they offer very cautiously, as a possible one.[2]

[1] *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, pp. 447 to 449.

[2] *Idem*, p. 449.

615. Application of this Mechanism to Twinning Lamellæ.—Extending this conception of the formation of a mechanical twin from the unit of Fig. 69 to the slices *S*8 to *S*14 which, in Fig. 68, form such a twin, the stress *S*8 at the bottom of slice 8 is supposed to rock each of the units of this slice about its right-hand side, corresponding to *vu* of Fig. 69, from the position shown in unbroken lines at the bottom of slice 8 into that shown in broken lines at its top. This common twisting of these units of slice 8 naturally raises and carries slice 9 and all at its left. At the same time, yielding to stress *S*9, all the units of this slice are rotating exactly like those of slice 8, thus bringing slice 9 into the position shown in broken lines. In like manner the twisting of the units of slice 9 has lifted slice 10 and all at its left. Thus the simultaneous rotation of the units in all the several slices lifts each slice a little more than the one next on its right, and the sum of these rotations creates the step $b'c'g$, thus changing the metal which takes part in the movement from its initial position *cghd* to the symmetrical or twinned position $c'ghd'$. At the same time the constituent particles of these various units rotate each about its center, from their initial direction shown in narrow unbroken lines at the bottom of the several slices to that shown in narrow broken ones at their top. Thus a mechanical twin forms.

This cumulative action by which each slice is raised not only by the rotation of its own units but also by that of the units of all the slices at its right, explains the twinning of the Neumann lamellæ by wheeling, with the formation of a marked step as shown in Figs. A to D of Plate 36. But instead we may imagine that the rotation of the units of each slice takes place not about their right-hand side but about their center, so that there is no translation. This would fit the theory to annealing twinning by facing about.

In the formation of a *slip band* shown in Fig. 67, as explained in §458, the rocking of the units of the several slices, though alike in all, is yet according to these authors only to the random position $c'ghd'$, and the initial orientation of the particles which make up those units is retained, as is indicated by the broken narrow lines, so that the condition differs from that of mechanical twinning in that not only the outward shape of the units and of the slices which they make up, but also the orientation of the constituents of those units, are out of equilibrium.

In the formation of an *annealing twin*, the final stage shown by Fig. 68 is reached through that in Fig. 67. First, during the deformation the outer form, both of the mass involved as a whole and of its various units, is twisted toward but not quite to the symmetrical or twinned shape, without change in the internal orientation of those units, that is without rotation of their constituent particles. Second, during the annealing the twisting of the outward form of both the whole mass and its various units is rectified into the symmetrical shape, and simultaneously the constituent particles of those units face about into the symmetrical orientation.

616. Discussion of the Twinning Mechanism of Osmond and Cartaud. Slip Bands Due not to Twinning but to Slip.—This theory is of two distinct parts, its explanation of twinning and that of slip. The first seems perfectly consistent with itself and with the facts in general. The explanation of twinning which I give in §611 may indeed be regarded as only an elaboration of it.

To the second part, explaining the slip bands as representing incomplete twinning, there is an obvious objection that the thickness of a Neumann band is of a far higher order than that of the region involved in the motion which causes a slip band, and that that of the annealing twins is of a still higher order. This objection may, I think, be set aside, because we may assume that these very able writers understood it, and assumed that their readers would understand that their mechanical twinning, resulting in the formation of a Neumann lamella, represented an aggregation of a very large number of motions of which each one by itself is the equivalent of that which causes one slip band, and that an annealing twin represented a corresponding greater aggregation. In short, in postulating similarity of mechanism they probably did not intend to imply identity of scale.

If this is conceded, it is indeed attractive thus to link slip bands and Neumann bands together as successive stages of a single general process, for on one hand they are identical in crystallographic orientation and in the shape of their profile, and on the other hand certain of their differences fit this theory. Thus the forcible deformation of all the crystal units in the region covered by slip, that is the inequilibrium in which they retain their internal orientation in spite of being forcibly twisted out of shape, might serve to explain why the plastic deformation which causes slip bands increases the hardness, volume, and solution pressure, and affects the other properties, while the symmetrical deformation which causes Neumann bands does not because it does not cause this inequilibrium.

But this theory does not, I think, explain satisfactorily why the slip bands differ from the Neumann bands in failing to reappear on repolishing and etching. Of course, the Neumann bands ought to reappear because the difference in orientation between them and the enclosing metal would naturally cause a difference in etching tint, but the slip bands also ought to, because the supposed inequilibrium of the region affected by slip ought to affect its etching tint.

To link the slip bands with annealing twinning by holding that both represent wheeling is even more difficult, because this form of twinning seems to be not by wheeling but by facing about.

Perhaps the strongest reason why slip is more probable than wheeling twinning as the mechanism of the deformation which slip bands represent, is that it explains more readily the greatness of the deformation, for obvious reasons. The difficulty with which traces of slip bands can be found on repolishing and etching, though this process develops the Neumann bands as very wide areas, for instance in Fig. F of Plate 34,

goes to show that the slip band strata, the strata involved in the formation of slip bands, are incomparably thinner than the Neumann lamellæ. The deformation which wheeling twinning can cause is clearly limited, and only of the order of the thickness of the layer which twins, whereas that which slip can cause may exceed indefinitely the thickness of the layer involved in the slip, the slip layer. For instance, the thickness of a lamella which is to cause the risers of Fig. E of Plate 25 by wheeling twinning must be of the order of magnitude of the height of those risers, and therefore wide enough to be extremely prominent in an etched section; whereas the thickness of one which is involved in the formation of those risers by means of slip might be only that of a few molecules, and hence invisibly thin.

If, now, the wheeling twinning of the Neumann lamellæ can cause only the extremely slight deformation which accompanies their formation, (§620), how could the incomparably greater deformation which accompanies the formation of the incomparably thinner slip band strata be caused by this same mechanism? Even allowing for any excess of the number of slip bands over that of Neumann bands, this consideration argues strongly that the supposed wheeling twinning of the slip band layers fails to explain the greatness of the deformation which may accompany their formation, whereas slip explains it readily.

These considerations are reinforced by those in §§621 and 622 tending to show that neither mechanical nor annealing twins are likely to contribute greatly to the plastic deformation.

617. Additional Reasons for Referring Slip Bands to Slip.—First, the abundant generation of amorphous metal to which the hardening and other marked changes caused by plastic deformation point, results more naturally from slip than from twinning. Slip would naturally generate amorphous metal along the extended surfaces of the slipping planes quite as polishing does on the outer surface (§§508, 509 and 512), and also through disregistry at the grain boundaries (§481). But such generation of amorphous metal by twinning ought to be confined to the grain boundaries (§619), that is to a region of a far lower order of magnitude. It is on the whole easier to refer these very pronounced hardening and like effects of deformation to amorphization along the far greater region in which it would occur if the mechanism were slip, than to that along the very limited region in which it would occur if the mechanism were twinning. Indeed, the fact that the hardening caused by cold rolling is almost as great in coarse as in fine-grained ferrite (Table 20B, lines 16 to 21), in spite of the great difference in their extent of boundary metal, shows that mere boundary amorphization cannot account for the hardening, which therefore refers itself to intragranular amorphization, and through this to slip, because twinning causes no hardening (§606).

Second, the progressive widening, under repetitive plastic deformation, (§422) of the bands which develop within the grains themselves, seems to

fit more smoothly with the conception that what is here occurring is slip along a great number of adjoining parallel slip planes, than with the alternative hypothesis that mechanical twins are forming here.

These two contributory reasons are less cogent than that already given in §616. These reasons collectively give us ground for inferring that the mechanism which causes the slip band is not twinning but either slip or fluid motion. In §682 we shall see reason for inferring that slip is by far the more important of these two, and hence that the slip bands are named aright.

The resemblance between the profiles of the most regular of the slip bands, Fig. E of Plate 25, and those of the Neumann bands, Figs. A to D of Plate 36, hardly carries weight, because such steps could result as well from slip as from twinning by wheeling.

This inference is all the stronger from the fact that it is in the relatively early stages of deformation that such high risers as those of Fig. E of Plate 25 form,[1] for the slip bands then formed are exactly the ones which refuse most stubbornly to reappear on repolishing. Yet if, as the twinning theory of slip bands implies, the width of the involved area is of the order of the height of these risers, it is great enough to be not only visible but prominent on repolishing and etching. Hence the failure of the early slip bands to reappear is all the harder to reconcile with the hypothesis that they represent wheeling twinning, though it tallies well with the slip hypothesis.

618. Summary of the Discussion.—I find no phenomenon of either twinning or slip which is simplified or more readily understood on this theory that slip and twinning are different stages of a single wheeling process, nor anything here to recommend this theory of slip. On the contrary the objection to it pointed out in §476, that it increases the difficulty of explaining the failure of slip bands to reappear on repolishing and etching, operates against thus identifying slip with either form of twinning, because this theory fails to explain the difference that whereas this reappearance is invariable in the case of twins, both mechanical and annealing, it is not in the case of slip bands.

Moreover, this wheeling theory of slip bands fails to explain the great deformation which accompanies their formation, whereas the slip theory explains it readily.

The facing-about process by which annealing twins seem to form differs sharply from the wheeling process by which slip bands form on this theory.

Thus I find no valid support here for the wheeling as distinguished from the slip theory of slip bands.

619. Twinning should not Generate Amorphous Metal Except at the Grain Boundaries.—The reasons which led us to infer that slip generates amorphous metal along the slip planes (§§508 and 509) do not apply to twinning, which lacks the rubbing action. We find no reason why twinning should generate amorphous metal, and no evidence that it does. The con-

[1] Rosenhain, *Journ. Iron and Steel Inst.*, 1906, No. II, vol. 70, p. 199.

stituent particles of the space lattice need not be in actual contact with each other, and the lattices as a whole may well be resilient enough to yield elastically to whatever momentary change of shape occurs in the twinning rotation itself, and to return elastically at the end of the rotation to an equilibrium symmetrical with the initial.[1] That twinning does not affect either the luster of the surface of calcite (§550), or the hardness of iron (§606), indicates that it does not disturb the structure, but only rocks it into a new and perfectly equivalent symmetrical position, for any disturbance would be likely to affect that bright luster, and almost certain to cause one of its most striking and constant effects, hardening. I am not aware that twinning, where it has been proved optically to occur, has been shown to cause any effects of this nature.

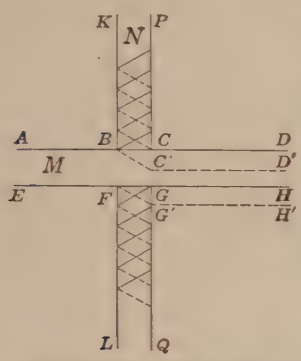

FIG. 106.—Movement of the enclosing metal implies faulting, but that of the lamella does not.

It is true that twinning by wheeling might rupture the grain boundaries (§613), because the very causes which lead to the disregistry of slip planes at those boundaries lead to that of twinning also. Here it is instructive to remember that the meteorite in which mechanical twins failed to change the hardness may well have consisted of a single grain, so that in spite of this case twinning may cause hardening along the grain boundaries. But the region here involved is very small compared with that along the surfaces of the slip planes within the grains, so that the hardening and other effects of amorphization, in the mass as a whole, should be much less in twinning than in slip.

620. Reciprocal Faulting[2] of Neumann Bands.—If M of Fig. 106 represents a Neumann band which has already formed, and N one which is about to form by the wheeling of its units from the position indicated by the unbroken oblique lines to that shown by the broken ones, this wheeling will evidently carry all the metal at the right of PQ along with it, faulting M by moving its right-hand half from $CDHG$ to $C'D'H'G'$. The distance covered by this throw will evidently be proportional to the thickness of N, and will vary with the angle which the direction of $N's$ motion makes with the plane of the section, being nil if that motion is normal to that plane.

Such faulting is shown in many unetched sections.[3] For instance, the

[1] Edwards seems to overlook these considerations (*Journ. Inst. of Metals*, 1915, No. 2, "Metallic Crystal Twinning by Direct Mechanical Strain," pp. 12–16 of preprint). See my remarks in the discussion of this paper.

[2] *Fault*, a severing of the continuity of a body of rock by a break through the mass, attended by movement on one side or the other of the break, so that what were once parts of one continuous stratum are now separated ("Century Dictionary and Cyclopedia," vol. III).

[3] Faulting of Neumann bands can be traced in micrographs shown by Osmond, Fremont, and Cartaud, *Rev. de Metallurgie, Mem.*, 1904, Vol. 1, Figs. 9 and 10, p. 20, and by Osmond and Cartaud, *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, Figs. 18 and 19, p. 464.

dark band *M* of Fig. B of Plate 32 is seen, on holding a straight-edge along its right-hand end, to be faulted by each of the successive bands of the dark *N* group. Part of the apparent curving of *P* is seen to consist of a series of minute faults caused by the *N* group, though this faulted line does not interpret itself confidently as a Neumann band rather than as part of a grain boundary. In Fig. *D* of Plate 32 the faint *F* series is faulted slightly by several members of the *G* series. The apparent cubic Neumann band *G* of Fig. B of Plate 35 appears to have been faulted twice by others running diagonally, the left-hand fault being parallel to a 211 intersection.

The proportion of cases in which faulting is not evident is not greater than can be explained by the narrowness of the bands[1] and the consequent slightness of their throw, jointly with the fact that it is only the component of the throw parallel to the plane of the microsection that can thus be disclosed.

In etched sections faulting is hardly to be looked for, because the furrowing by the etching reagent causes an increase in the apparent width of the Neumann band which is of a higher order of magnitude than the true width, and hence than the maximum possible throw by wheeling. This widening is shown for instance in Fig. E of Plate 34, in which very narrow Neumann bands seem to have been etched out spottily to a width of about $\frac{1}{8}$ in. In Fig. A of Plate 34, too, the longer etching has opened up a set of dots and furrows about $\frac{1}{8}$ in. wide, though the Neumann band which ought to exist between successive dots running in any given straight line is too narrow to be seen here. If visibly wide it ought to be seen here, because this figure represents a fracture, and not a polished surface on which the flow might conceal a Neumann lamella.

It is thus that we naturally explain the complete lack of apparent faulting in Figs. A and F of Plate 34. In each of these the number of directions is so great that some one at least of the sets of Neumann bands must have faulted another. On this magnification the very slight throw which the narrow Neumann bands could cause is masked by the furrowing caused by the etching. Conversely, the absence of visible faulting argues that the Neumann lamellæ here must be extremely narrow, in spite of their great width after etching.

Yet on the etched sections *D* and *E* of Plate 33, and *D* of Plate 34 where two Neumann bands (?) intersect, one, *M*, is clearly continuous and the other, *N*, as clearly discontinuous. More study is evidently needed. Thus there is the possibility that the etching bands in Figs. A and F of Plate 34 represent not true Neumann but only 211 partings (§590).

621. The Formation of Neumann Lamellæ is not an Important Cause of Plastic Deformation.—This is in general accord with the law covered

[1] The width of the steps in Figs. B to D of Plate 36 at first seems inconsistent with this alleged thinness; but here the magnification is about 24 times that of most of these micrographs. The width of band G of Fig. B, Plate 35, is explained by the great magnification. This width corresponds to the horizontal component of the risers of Figs. B to D of Plate 36.

by §584, that they form the more readily the more brittle the iron and the more rapid the deformation, because these are just the conditions which lead to the least total deformation before rupture. Their abundance in the extremely brittle bismuth tallies with this. Note how abundant and prominent they are in the extremely brittle silicon steel in Plate 32.

622. Annealing twinning probably does not contribute greatly to plastic deformation, as we may infer from our previous inferences (§613), that it occurs by facing about, or rather from the evidence from which that inference is drawn, and especially from the fact that the consummation of this twinning by annealing does not disturb even the previous high polish of the surface (§567). It is true that this consummation of the twinning brought about by the annealing is preceded by the deformation which sets up the twinning tendency, and the fact that this consummation causes no disturbance does not prove that the prior deformation caused none. But if, as we naturally infer, the consummation consists of two fractions, the rotation of the component parts of the various crystal units into their new twinned orientation and the accurate rectification of the twisting of the units themselves into their twinned outline, so that this rectification is a fraction of the prior deformation, we naturally expect it to be a random fraction, and therefore often a considerable one. Because this probably considerable fraction of the total deformation is so insignificant that it does not suffice to disturb even a previous high surface polish, we infer that the whole, of which it is such a fraction, does not cause great deformation.

622A. The Important Mechanism of Plastic Deformation Reduced by Elimination to Slip and Fluid Motion.—Because intergranular movement seems to be relatively slight and confined chiefly to the earlier stages of deformation, this present elimination of both mechanical and annealing twins as important causes of deformation leaves slip and fluid motion as the probable chief mechanisms.

CHAPTER 25

THE X BANDS

623. Summary.—What I call X bands are rough striæ, shown in Figs. G of Plate 36 and A to G of Plate 37, found on polishing and etching iron which has been severely deformed plastically. They resemble the lines found under these same conditions which Rosenhain interprets as the traces of the slip planes, in that rest and heating tend to prevent the etching from developing them. They are not interpreted readily as representing either the amorphous contact metal caused by the deformation, or twinned lamellæ. Their nature is not known.

624. Their Occurrence.—When ferrite, in the form of low-carbon steel, is deformed so as to cause slip bands, if it is next repolished so as to efface those slip bands and then etched, bands may appear at least in some of the grains recalling the slip bands strikingly. These bands do not appear on etching specimens which have not been deformed thus, and hence they must be connected with the deformation. Till we learn what they are we may call them "X bands." They are shown in Plate 37, especially in the grains there numbered 1, 2, and 3. Note how closely the X bands on grain 1, Fig. E, recall the slip bands of Figs. D and E of Plate 23.

TABLE 23.—INFLUENCE OF REST AND HEATING AFTER COMPRESSION ON THE BRINELL HARDNESS

Compression, lb. per sq. in.	Original	20 min.	5 to 6 hr.	24 to 25 hr.	38 days	24 hr. cold + 100° for 15 min.	24 hr. cold + 300° for 15 min.	29 days cold + 500° for 15 min.	29 days cold + 550° for 15 min.	29 days cold + 600° for 15 min.
	A.R.M. steel, No. 0, carbon 0.01									
21,200 first series	75	71	71	75	81	84	87	.....	.....	.....
21,200 second series	75	75	75	77	81	81	84	.....	.....	.....
30,300 first series	75	84	81	84	86	95	99	84	84	81
30,300 second series	75	84	81	84	86	91	94	87	88	87
	Steel No. 17.1, carbon 0.147, after heating to 800°, etc.									
31,700	130	118	114	124	138	137	140	.....	.....	.....
43,000	130	126	124	124	140	137	140	149	149	140
67,900	126	140	146	134	179	170	163	.....	163	163
90,500	126	163	149	163	196	179	179	187	183 179	174 174

Figs. A to E of Plate 37 represent the specimens used in the tests of Figs.

78 and 79, p. 365, and Table 23, of the very pure A.R.M. steel, with only 0.01 per cent. of carbon. Figs. G of Plate 36 and F and G of Plate 37 show the same steel after punching, followed in the last two cases by reheating.

The most striking bands are those shown in Fig. F of Plate 37. Their appearance is much the same when the heating which this specimen underwent is omitted, Fig. G of Plate 36. They are rough striæ which in general follow the same direction in any one grain, but change direction sharply at the grain boundaries.

625. Rosenhain's Observations.—He finds[1] such bands on repolishing and etching, though unfortunately he gives no micrographs, and his description leaves us in some doubt whether his bands are identical with what I show in Plate 37. He finds that they coincide in position with the slip bands, that is, with the steps on the previously polished surface which the repolishing has effaced, and that they are the less persistent the slighter the deformation which has caused them.

Thus if the deformation has been slight his bands appear on etching only in case all heating, even that incidental to the usual dry polishing, is avoided, and only in case the etching occurs very soon after the deformation. Thus either rest or slight heating suffices to prevent their appearing.

If, on the other hand, the deformation has been severe, then neither rest nor gentle heating prevents the appearance of these bands on etching. Unfortunately, he gives no indication of how high the temperature may go without preventing the appearance of these bands on later etching.

He further finds that in very pure iron the persistency of these bands is especially brief, covering only a few hours.[2] This probably refers to cases of very slight deformation. A micrograph given me by Stead shows curved markings recalling the X bands, in a bar of nearly pure iron cooled very slowly from 1,300°, polished and etched. He refers the curvature to unequal contraction.[3]

626. My observations are represented by the data in Table 23 and Figs. 78 and 79, in addition to the micrographs of Plate 37 already referred to. Be it understood that the presence or absence of X bands was determined in each case by polishing and etching after the heating or rest. The more important results are as follows:

[1] Rosenhain, *Engineering*, 1913, vol. 96, p. 510. In his "Introduction to Physical Metallurgy," 1912, he shows, Fig. 132, Plate XXIX, in steel which has been deformed by a cutting tool, bands which resemble those of Plate 37. But I do not find any indication as to whether these are the bands which, in his opinion, represent the amorphous metal formed along the slip planes. X bands are shown also by Brearley in the necked region of a tensile test piece which elongated 49 per cent. in 2 in., and are interpreted as the traces of slip bands. ("The Use of Microscopic Methods," *Sheffield Soc. of Engineers and Metallurgists*, 1909–1910, pp. 17–18, Fig. 56.) What he shows in his Fig. 54 may possibly have been developed after repolishing and etching, that is may be X bands, but the context does not make this clear. What at first might be taken for X bands in Heyn's Fig. 52, Martens-Heyn, II, A, "Materialenkunde," 1912, p. 223, and Plate 11, are slip bands formed on a polished and etched face, that is on one etched before and not after their creation.

[2] *Journ. Iron and Steel Inst.*, 1906, No. II, vol. 70, p. 207.

[3] Private communication, Dec. 5, 1914.

Rest, even as long as 38 days, after deformation still leaves abundant X bands, as in Fig. C of Plate 37.

Heating progressively weakens these bands, and the more effectually the less the deformation has been. Thus while Fig. D, taken after heating to 100° for 15 minutes, shows abundant bands, Fig. E taken after heating for 15 minutes to 300°, 24 hours after the compression, shows much fewer, though the compression was the same in both cases. After the extreme deformation of punching, the bands are abundant and pronounced even after heating to 600° for 15 minutes, as in Fig. G of Plate 37. Yet after heating for 15 minutes to 800° these bands could no longer be developed.

After the slighter deformation caused by a pressure of 21,000 lb. per square inch these bands were fainter, and their appearance was much more effectively prevented by heating, than after the pressure of 30,000 represented in Figs. A to E of Plate 37. For instance, after exposure for 15 minutes to 300° few distinct X bands could be found.

Thus whether the X bands which I find and those to which Rosenhain testifies are or are not the same, they are alike in increasing in prominence and persistency with the degree of deformation, and in being weakened or even prevented from appearing at all by gentle heating.

In many fields, for instance those of Figs. A and D of Plate 37, the larger grains are filled with X bands, but the smaller ones are wholly free from them. There is much to suggest that these small grains, for instance those of Fig. A, have already undergone the early stages of the metamorphism which occurs in plastically deformed steel when it is reheated. This metamorphism consists first in the complete break-up of the old grains which have been distorted by the deformation, probably into an enormous number of submicroscopic new grains, and second in the gradual growth of these new grains.[1]

A grain selvage, like those which accompany the slip bands on polished but unetched surfaces is seen from *a* to *b*, and from *c* to *d* in Fig. F. These selvages were very marked in specimens heated to 800° after great deformation.

627. What are the X Bands?—They might be the traces of slip and thus represent the slip bands. If, as is currently held, these are only the surface steps or uplifts at the outcrops of the planes along which slip has occurred; and if, as Beilby and many others believe, the friction of this slip along these planes makes a thin contact layer of metal along these planes amorphous, much as the friction of the polishing reagent certainly makes a thin layer of a surface which we polish amorphous; then any new plane cut across these slip planes would necessarily cut across this contact layer of amorphous metal at each slip plane, and this would be true of the new plane exposed by repolishing. Because this amorphous metal has greater solution pressure

[1] See Heyn, Martens-Heyn, II, A, "Materialenkunde," 1912, pp. 233–235. Also Chappell, *Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, p. 460, where a very full and interesting discussion of this recrystallization is given.

than the crystalline metal surrounding it, it might well etch differently from that crystalline metal, and thus show on such a repolished and etched surface as a band like in position and direction to the slip bands. Hence at first sight it is most natural to interpret these markings as representing this amorphous metal, thus corresponding to the slip bands themselves, and being in effect "slip traces."

But the slip in addition to causing these amorphous layers may also cause mechanically twinned layers, and these may well run parallel to the slip planes themselves. Because such twinned layers habitually etch differently from the enclosing and differently oriented metal, they too might well appear on etching the repolished surface as bands like in position and direction to the slip bands. Hence these X bands may be twinned layers, or at least layers of different even if not of symmetrically different orientation.

In order to learn whether the X bands represent "slip traces," that is to say metal made amorphous along the contacts of the slipping planes, we should ask whether rest and heating affect them as they do this amorphous metal. Because the amorphous metal is assumed to be the cause of the hardening which plastic deformation causes, we may answer this question by asking whether rest and heating affect the X bands as they affect the hardness of the plastically deformed metal. To this end I have made the investigations which are described in the following section.

628. Influence of Rest and Heating on the Hardness Caused by Plastic Deformation.—I showed in the first edition of this work[1] by abundant cited evidence that rest and moderate heating intensify the strengthening effects of plastic deformation, but that heating to redness effaces them, and Muir[2] later re-established this effect of rest and gentle heating in the most striking way.

Because rest and heating thus have generally like effects, the effect of heating may be overlooked, indeed it may not occur, if there has meanwhile been a long rest, for instance after wire-drawing or other form of plastic deformation, the rest in that case anticipating the heating. It remains to be shown how complete this anticipation is, and whether heating after rest has this strengthening effect, in short whether heating and rest act cumulatively or as alternatives. Experiments which I have tried tend to show that this anticipatory effect of rest may be complete.[3] This may

[1] "The Metallurgy of Steel," Scientific Publishing Co., N. Y., 1890, p. 212. This striking fact seemed to me so important that for many years I had my students carry out experiments to illustrate it. See my "Metallurgical Laboratory Notes," Boston Testing Laboratories, 1902, p. 29.

[2] "On the Recovery of Iron from Overstrain," *Phil. Trans. Royal Soc.*, 1900, A, vol. 193, p. 1.

[3] A wad of steel, 38 days after plastic compression, was slightly softened by heating to 100°.

A series of tests, made at my suggestion by an important establishment which draws shafting cold on a very great scale, did not show that heating cold-drawn steel to 300° affected it materially, perhaps for this reason.

There is much in the present evidence to suggest that, though heating thus increases the Brinell hardness and the elastic limit, which are connected with the cohesion existing before any

explain why Professor Goerens'[1] very extended and valuable researches on plastic deformation did not detect this strengthening effect of heating, but instead showed a slight weakening. This will be considered more fully in another volume.

In order to show afresh this strengthening effect of gentle heating and rest, I had short wads of nearly pure iron compressed, while standing one on another, so that all were exposed to identical pressure under nearly identical conditions, and I thus avoided the later effects which would have been caused by machining the specimens to shape after the compression.[2]

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large degree of plastic deformation occurs, it has much less effect on the tensile strength, which represents the cohesion after great plastic deformation. This in turn suggests that here, too, the heating only anticipates and does not add to the strengthening which occurs between passing the elastic limit and reaching the maximum load, called the tensile strength.

[1] *Carnegie Memoirs, Iron and Steel Inst.*, 1911, vol. III, p. 379, and especially p. 399.

[2] The materials tested were the very pure A.R.M. steel of 0.01 per cent. of carbon and the No. 17.1 steel of 0.147 per cent. of carbon (see Table 39, p. 616). The latter had been drawn cold, and was first heated to 800° so as to efface the effects of the cold drawing, and furnace cooled. Wads, ¾ in. in diameter and ¾ in. thick, were then cut from both, and annealed by heating to 600° and furnace cooling. They were then smoothed so as to be ready for the Brinell test.

These wads, in lots of four, were next set on each other so as to form little columns, and these columns were compressed endwise, one at a time, between the plates of a tensile testing machine.

In each case the pressure was raised steadily till the maximum load for that experiment was reached, when it was at once released.

One wad of each set was then used for a rest experiment and one or more for the heatings. The deformation along those surfaces of the upper and lower wads which were in immediate contact with the plates of the press must have been lessened by the friction, in spite of their being previously greased. Hence these upper and lower wads were not used for either the rest or the more important heating tests, being used only for the 500°, 550°, and 600° heatings.

The 100° heatings were done in water boiling vigorously, the wads having been put in the water while it was still cold. The other heatings were done in a tube heated very rapidly by electric resistance in the usual way, and in an atmosphere of either hydrogen or carbonic oxide.

In the rest experiments the hardness was determined about 20 minutes after releasing the pressure, again after between 5 and 6 and between 24 and 25 hours, and again after 38 days after that release. Each determination was made at a new spot on the wad.

The heating experiments to 100° and to 300° were made between 24 and 25 hours, and those to 500°, 550°, and 600° were made 29 days after the release of the pressure. In all cases the high temperature was maintained for 15 minutes.

The conditions are not faultless. The ratio of the length to the diameter of the little columns may have been great enough to cause some unevenness of pressure through incipient buckling.

In extending this experiment I purpose to give uniform deformation by rolling a plate cold. This will give specimens so large as not only to permit far more numerous hardness tests, but also to reduce thus the influence of local variations in hardness from segregation or other causes. It is not thought that any material error has arisen from segregation, because the hardness determinations made on the 17.1 steel after annealing at 800° and before any later treatment gave very concordant results. The hardness of the first lot of 12 wads, which had been heated to 800° for 45 minutes, varied only between 128 and 131. That of the second lot of six which had been heated to 800° for 2½ hours, was identical, 126. The specific gravity of six of the first lot of wads varied between 7.8388 and 7.8394; that of four wads of the second lot varied only between 7.8394 and 7.8396. The density was determined after many of the tests, but I must defer the discussion of the results. These specific gravities refer to 15°C. for both steel and water.

Two lots of the A.R.M. steel, both of which underwent a compression of 30,000 lb. per square inch, gave identical hardness results for each series at each of the five stages of the tests, as shown in Fig. 79.

Though the data are not abundant enough for great quantitative accuracy, they suffice in my opinion to establish provisionally the laws here deduced.

I then had the Brinell hardness of these wads determined after various periods of rest and after exposures to various temperatures, with the results shown in Table 22 and graphically in Figs. 78 and 79, p. 365.

629. Influence of the Degree of Deformation on the Brinell Hardness Determined Immediately Afterward, Fig. 78.—A determination of the hardness 20 minutes after the release of the compression shows that though a moderate compression may soften the metal slightly, a greater compression hardens it, and that every increment in pressure, beyond the lowest used, leads to an increment of hardness. This is true of each of the two steels tested.

630. Influence of Rest after Deformation, Figs. 78 and 79.—After 38 days the hardness was in every case greater not only than the original but than that found 20 minutes after the compression. The hardness of each of the varieties of steel increased with the severity of the compression, though not proportionally to it. After the severe deformation caused by a compression of 67,900 or 90,500 lb. per square inch the hardness increased on rest very greatly, the 38-day rest in one case causing an increase of hardness thrice as great as that caused by the deformation plus 20 minutes' rest.

Yet though rest thus increases one of the effects of plastic deformation, its hardening, it decreases another and simultaneous effect, the increase of volume or loss of density. Thus Lea and Thomas[1] find that a rest of from 35 to 38 days, strangely enough almost exactly the maximum period of rest in my experiments, causes an increase of specific gravity of between 0.00243 and 0.00426, or of between 0.031 and 0.054 per cent. in the case of mild steel. The compressive stresses which they used were far greater than mine, from 130 to 198 tons per square inch (291,200 to 443,520 lb. per square inch).

After periods of rest between 20 minutes and 24 hours, Fig. 79, the changes in hardness are complex. For each material the slighter compressions are followed either by no change in hardness or by a softening up to the end of the 5–6-hour or even of the 24–25-hour period. After the greater compressions a gain of hardness at the end of the 20-minute period was in each case followed by softening at a later period.

In seven out of the eight cases there is both a hardening at some period during the first 24 hours and a softening at some other period, while in the eighth case the initial hardness remained unchanged at the end of 20 minutes and 5–6 hours, and increased slightly from then on to 24 hours.

631. Influence of Gentle Heating after Plastic Deformation on the Brinell Hardness.—Heating to 100°[2] increases the hardness of the plas-

[1] "Change in Density of Mild Steel Strained in Compression beyond the Yield Point," *Engineering*, 1915, vol. 100, p. 1 (July 2). These interesting determinations are of the influence rather of a given strain than of a given stress. It was found, as might have been expected, that the effect of the compression varied greatly with the length of time during which the specimen was loaded, and hence the time was kept constant.

[2] It should be noted that this heating to 100° took place 24 hours after the compression, and that then, before this heating occurred, the hardness differed materially from that found immediately after the deformation.

tically deformed steel in every case. For each kind of steel the gain in hardness by this heating, over the hardness at the time immediately preceding the heating, increased with the severity of the deformation which the metal has undergone.

Heating to 300° causes a further increase of hardness beyond that found after heating to 100° in the six cases in which the compression was less than 67,000 lb. per square inch, but either no change or a slight softening in the two cases in which the load was greater than 67,000 lb. In no case did it reduce the hardness to that found 20 minutes after the deformation.

Heating to 500°, 550°, and 600° usually reduced the hardness to much less than it was after heating to 300°, but it never removed the whole of the hardening caused by the deformation, nor left the metal as soft as it was 20 minutes after the deformation.

632. Parallel Effects on Tin.—Rose finds that cold-rolled tin anneals itself and softens spontaneously at any temperature between 10° and 100°; that this softening does not complete itself at the room temperature even in 97 days, but that it is "practically complete . . . . in less than a minute at 100°."[1] There is no indication that this softening is temporary, as in the case of iron, to be replaced by hardening later.

633. Plural Proximate Effects of Plastic Deformation.—Here we have very striking complexities in the behavior of plastically deformed ferrite. The invariable hardening which occurred between the first and the thirty-eighth day after the compression contrasts strongly with the almost invariable softening which occurred at some time during that first day, the only exception being one case in which the hardness showed absolutely no effect of the deformation for 5–6 hours and then increased slightly. This contrast tallies with Muir's observation that though immediately after the deformation the elastic limit was apparently zero, the stress-strain diagram curving from the origin, yet on rest or gentle heating the elastic limit rose progressively to far above the initial.[2]

These complexities suggest that the compression causes two conflicting effects, which according to the special conditions of each case may lead to a net softening, a net hardening, or an exact balance, but that as time goes on the hardening effect outweighs the softening one. This inference receives some support from the fact that whereas rest exaggerates the effect of deformation as regards hardness, it lessens that effect as regards density.

634. Do the X Bands Represent Amorphous Iron?—Rosenhain believes that they do; that they are the intersections of the plane of the microsection with those slipping planes the slip along which has caused the slip bands, and hence that they are to a section re-polished and etched after deformation what the slip bands are to a section polished before deformation. Yet I do not find any links by means of which he connects the two.

[1] *Trans. Faraday Soc.*, 1915, vol. 10, May, p. 282; and *Journ. Inst. Metals*, 1915, No. 2, in paper of Edwards on "Metallic Crystal Twinning."

[2] Muir, *Phil. Trans. Royal Soc.*, 1900, A, vol. 193, pp. 9 and 22, and 1902, A, vol. 198, p. 1.

Let us see what light the evidence which I have now given throws on this question. We have just seen that, after plastic deformation, though rest and gentle heating weaken or efface the X bands in the sense that etching now develops either fainter ones or none at all, yet they have the opposite effect on the hardness which that same deformation has caused, increasing instead of lessening it.

Two explanations offer themselves. The plastic deformation may have plural immediate effects, of which one leads to the etching differences and another to the hardening. Rest and heating may lessen the former effect while exaggerating the latter. Or the heating and rest, while returning part of the amorphous metal to or toward the crystalline state and thus lessening the etching and density effects, may simultaneously bring the remaining amorphous metal to a stronger state as in the setting of cement, and thus exaggerate the hardening. We know so little about the properties of the amorphous metal that this second explanation is hardly improbable.

635. Do the X Bands Represent Twins?—For like reason we are reluctant to refer the X bands to twinning. We see no reason why gentle heating should efface true twins by causing them to revert to the orientation of the enclosing metal. We have seen in §585 that Neumann bands in ferrite certainly persist through far higher heating.

Were it true that X bands caused by slight deformation are prevented by rest and gentle heating from appearing, and that those caused by more severe deformation are not prevented unless the heating is pushed very high, for instance to $Ac3$, so that they differ in kind, the former being effaceable and the latter persistent, we might suppose that the effaceable ones represent zones which had changed orientation but without reaching the twinned position, and hence are unstable and tend to revert, but that the persistent ones represent true twins, and hence are stable. But the difficulty in the way of this explanation is that the effaceableness seems to vary continuously with the degree of deformation. After undergoing a pressure of 21,000 lb. per square inch slip traces disappear on heating to 300°; if this pressure is 30,000 lb. they become less marked at 300° but do not disappear; whereas those caused by the great deformation of punching are very prominent after heating to 200° or 300°, and rather prominent even after heating to 600°, disappearing almost completely after heating to 800°.

636. Evidence Given by Invar.—A further suggestion that these X bands do not represent twinned layers, formed along the slip planes simultaneously with the formation of slip bands on the polished surface, is given by the fact that when both annealing twins and slip bands have been developed abundantly in invar, repolishing and etching, though it develops the twins prominently, develops only few and faint X bands. Were these X bands twins, then we should expect them to reappear prominently in this alloy because it twins so very readily and strikingly.

In short, the nature of these X bands is yet to be discovered.

CHAPTER 26

THE DEFORMATION LINES IN HADFIELD'S AUSTENITIC MANGANESE STEEL[1]

637. Summary.—The lines which form when this substance is deformed may be divided into the "surface bands," which form on previously polished surfaces, and the "etching bands," which are developed by polishing and etching after the deformation. These may be further divided into those found when there has been heating after the deformation and those found when there has not. In addition to these normal etching bands certain special eutectiform bands are sometimes developed by etching. These are interpreted provisionally as representing a eutectic.

The surface bands I interpret as slip bands. The etching bands after slow cooling contain cementite. Those found in unheated specimens seem to be related to the surface bands much as the X bands of ferrite are related to its slip bands. Their shape recalls the Neumann bands, while the analogy of niccoliferous austenite and the great hardening which accompanies their early formation point to beta-ization in addition to amorphization.

Thus there are suggestions of eight distinct entities: slip bands, annealing twins, Neumann bands, cementite partings, X bands, amorphous iron, martensite, and the eutectiform mass.

638. Introductory.—Hadfield's manganese steel, an alloy of about 11 to 13 per cent. of manganese and from 1.15 to 1.50 per cent. of carbon, in its usual industrial austenitic or gamma state brought about by quenching it in water from about 1,100°, differs from most metals and alloys as regards its behavior under deformation both in developing annealing twins only with difficulty and in developing very easily certain prominent lines resembling slip bands in form but differing from them in the greater persistency with which they reappear on repolishing and etching, even after high heating. Our interpretation of the nature of these lines may well be deferred till we have studied their life history, which we will take up at some length after noting the annealing twinning of this alloy. Meanwhile they may be called simply "the manganese steel lines."

These lines as they appear on the previously polished surface of this quenched alloy after it has been deformed, but without etching, are shown in Figs. A and B of Plate 29, B, E, H and K of Plate 38, and D, E and G of Plate 39; in the same alloy after polishing and etching in Figs. G of Plate 38 and A, B, C, F, H and I of Plate 39; and after reheating and then repolishing

[1] Part of this chapter is taken from a paper by the Author and A. G. Levy, "Are the Deformation Lines in Manganese Steel Twins or Slip Bands?" *Bulletin Amer. Inst. Min. Eng.*, April, 1915. My further study of the subject has led me to modify some of the inferences given there.

and etching in Figs. F of Plate 16, A and D of Plate 38, and E of Plate 42, in which they are more prominent and often thicker than under other conditions.

With the exception of the 33 per cent. thermit manganese steel referred to once in §639, the manganese steel discussed in this chapter is the normal Hadfield alloy of about 12 per cent. of manganese and 1.25 per cent. of carbon, quenched from 1,000° or 1,100° in water to bring it to the austenitic or gamma state, and without further thermal treatment except where the contrary is clearly indicated.[1]

639. Annealing twins of the common broad type, Figs. A and B of Plate 29, can be developed in this alloy in its cast and water-toughened state by the usual course of overstraining, annealing, polishing, and again straining (§559). But their development calls for a very great degree of initial deformation, a fact which has given rise to the belief that this alloy does not twin at all in this way. Rolled specimens twin much more easily than cast.[2]

The twins in Fig. B of Plate 29 have the familiar staircase effect so well shown in Fig. F of that plate, the lines which run lengthwise of the staircase having a common direction in the four odd-numbered zones and another common direction in the three even-numbered ones. The twinned zone in Fig. A of Plate 29 is typically parallel sided. These and many like twins were developed close to the fracture of a tensile test piece and therefore where the deformation was extreme, but none were found after a moderate deformation such as sets up the twinning tendency in most metals.

The difficulty of developing twins in this alloy is the more surprising because of the ease with which they are developed in some analogous alloys, 33 per cent. manganese steel nearly free from carbon, 25 per cent. nickel steel, and the 36 per cent. nickel steel called "invar." In the first of these, which in its cast state is composed of at least two distinct constituents and is twinless, I found typical twinned areas after quenching it from 1,100° without subsequent annealing. They are therefore mechanical twins. The invar of Fig. F of Plate 29 differs from Hadfield's alloy only in being niccoliferous instead of manganiferous austenite, and in lacking the 17 to 23 per cent. of dissolved cementite which the great carbon content of the former implies.

640. The Manganese Steel Lines.—The progress of their development, as traced on the previously polished surface of a wedge gently compressed in the manner described in §413 is shown in the middle column of Plate 38. For comparison the corresponding development of slip bands on the surface of a wedge of copper is shown in the right-hand column of that plate.

At the beginning of the deformation, shown in Fig. K, there are a few parallel lines, a gentle undulating disturbance of the whole surface, and the

[1] This alloy was kindly given by Mr. Knox Taylor, President of the Taylor-Wharton Iron and Steel Company. The thermit manganese steel was kindly made for this work by Mr. deCourcy B. Browne of the Goldschmidt Thermit Company of New York.

[2] J. H. Hall, private communication, May 3, 1915.

outlining of some of the grain boundaries, all three phenomena reproducing closely what occurs in other metals. Then these parallel lines become more marked, and are often crossed by a conjugate set, as in Fig. H. Next they become both more numerous and more prominent, as in Fig. E. Under still greater deformation the surface distortion or ruffling becomes so great that it tends to mask these lines, as in Fig. B. Thus parts of Fig. B are necessarily so far out of focus that they here look plane.

The lines in manganese steel and the slip bands in copper become visible at about the region where the undulation of the surface can first be recognized, and hence under about the same degree of deformation. Judged by means of the accompanying degree of undulation, the manganese steel lines appear if anything a little earlier than the slip bands in copper. Even where the undulation is very marked there are undulated areas in which no lines can be traced under a magnification of 50 diameters, and but few under a magnification of 500 diameters. Nevertheless it is possible that the whole of the undulation may be accompanied and accounted for by slip bands so minute as to escape detection.

641. Their Characteristics.—Whether caused by shock or by quiescent deformation they are in parallel sets, sometimes in one direction only in each grain, as in Figs. A of Plate 39 and K of Plate 38, but more often in two directions if on previously polished surfaces, as in Figs. B, E, and H of Plate 38 and C, D, E, and G of Plate 39, or even in three directions, as in the etched sections of Figs. F of Plate 16 and B of Plate 39. The direction of lines which appear to correspond to those of Fig. B of Plate 39 were identified by Osmond and Cartaud as octahedral.[1] These lines often recall martensite, when, as in Figs. F of Plate 16 and B of Plate 39, two sets of them are of about equal prominence. In their incipiency they may run but part way across a grain, as in Fig. K of Plate 38, but under greater deformation they usually extend completely across the grain in which they occur, as in Figs. F of Plate 16 and A of Plate 39. Though on reaching a grain boundary they generally stop, as in the last two figures, they may continue across grain boundaries, sometimes without changing direction, as in Fig. A of Plate 38. They may cross the dendrites without the least deflection, as in Fig. G of Plate 39.

They are usually nearly straight or curved smoothly and gently, but occasionally they are irregular, as in grain 1 of Fig. A of Plate 39, and as there they may even recall lamellar pearlite.

They very rarely fault each other sharply as slip bands in other metals often do, perhaps because for given deformation they are so numerous that the fault at any one intersection is so slight as to escape detection. Indeed, some at least of what seem at first to be faults on closer examination seem rather to be twinned areas. Thus at *e* in Fig. C of Plate 29 the strong line running N-80°-W at first seems to fault the line running N-25°-E which it intersects here. But closer examination shows that a like deflection occurs in the other N-25°-E lines where they are not intersected by the N-60°-W

[1] *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, p. 460.

lines, and that these deflections seem not to be faults but rather to represent a twinned area running from *a* to *b*. Then we see a like band running from *c* to *d*.

A very slight deformation suffices to cause them. Thus, though they are lacking, at least usually, on the etched surfaces of castings as taken from the sand, they may be made extremely prominent in these by the slight unintentional deformation which occurs when a casting, cut almost completely through, is then broken in two.

642. Hardening and Magnetization.—As they form under increasing deformation the metal becomes very slightly ferro-magnetic[1] and increases greatly in hardness. Thus the Brinell hardness close to a tensile fracture was 340,[2] whereas in the undeformed part of the same test piece it was only 217, so that there was a gain of 123 Brinell numbers. Hadfield and Hopkinson found a gain of about 320 Brinell numbers,[3] or five times that which I found in the tensile rupture of carbon steel of 0.11 per cent. of carbon[4] (Table 25, p. 466).

A somewhat similar form of austenite, that of Carpenter, Hadfield, and Longmuir, with nickel 19.91, carbon 0.41, and manganese 0.96 per cent., behaves in like manner. It is soft and nonmagnetic in its undeformed cast state, but on deformation becomes extremely hard, martensitic, and magnetic.[5]

This magnetization indicates that some of the metal is transferred to the alpha state. We may suppose that this transfer is a result of the amorphization which occurs along the slip planes. But the hardening of manganese steel caused by the deformation is so much greater than is readily explained by the very small quantity of amorphous alpha iron which the slight gain in magnetism calls for, that we naturally assume that much amorphous gamma iron forms simultaneously, and that this amorphization hardens gamma iron as it does the other malleable metals in general. Beyond this, we refer a large part of this hardening to the martensitization which the deformation causes, because this hardening is so much greater than that which occurs in other metals. This martensitization is proved by Carpenter,

[1] Hadfield and Hopkinson report that breaking this material always makes it "very slightly magnetic, perhaps of the order of ½ per cent. of that of pure iron" (*Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, p. 132). The hardening itself reminds us of Guillet's observation in 1906 that austenitic steels which in composition are near the line which divides the austenitic from the martensitic are prone to martensitize on either mechanical of thermal disturbance. That is to say, the austenitic state of all such steels is very unstable and is degraded easily into the martensitic ("The Industrial Future of Special Steels," *The Iron and Steel Magazine*, 1906, vol. 11, p. 92). Compare also Guillet, "Les Aciers Speciaux," Dunod, Paris, 1904, p. 132, and the Author, U. S. Patent No. 1,222,193, Dec. 29, 1914.. See note 2, p. 182.

[2] The surface on which the hardness was determined was a transverse section made by grinding the fracture smooth, and was thus as near as practicable to the region of fracture itself.

[3] They found the Brinell hardness of a tensile test piece after rupture 540, against about 220 in like material before straining (*op. cit.*, 1914, No. 1, pp. 112 and 124, and *Trans. Amer. Inst. Min. Eng.*, 1914, vol. 50, p. 486).

[4] "Are the Effects of Simple Overstrain Monotropic?" *Proc. Amer. Soc. Testing Materials*, 1914, vol. 14, p. 28. Here the greatest hardness found, close to the tensile fracture, was 195, against 132 at the grip-end where there was little if any deformation.

[5] *Seventh Report Alloys Research Committee, Inst. Mech. Eng.*, 1905, pp. 949, 953.

Hadfield, and Longmuir, whose micrographs show clearly that the normal austenitic microstructure of their 19.91 per cent. nickel steel becomes martensitic on deformation.[1] That the hardening exceeds that of other metals under like conditions is shown in part by the Hadfield and Hopkinson increase of 145 per cent. in tensile rupture (Table 25), but more convincingly by the behavior of such austenite under the hack saw. Carpenter, Hadfield, and Longmuir record an observation like that which forces itself on all of us who experiment with such material, that their niccoliferous austenite, though cut rather easily at first by a hack saw, refuses stubbornly to yield after the cut has become about ⅛ in. deep. Replacing the blade thus dulled with another enables one to cut a little deeper[2] (§644).

If we accept my estimate that the Brinell hardness of this steel before deformation is less than 125 (§644), its raising to 540 or to more than fourfold in the Hadfield and Hopkinson test exceeds the hardening of other metals by deformation so far as to be referred in large part to the beta iron implied by the martensitization, especially because the possible deformation is limited by the accompanying brittleness.

Another possible cause of the hardening is the precipitation of cementite caused by this amorphization, as is explained in §653.

643. The chief reasons for the usefulness of this alloy are probably its ability not only to retain continuity but to strengthen through martensitization, and the ease with which this martensitization occurs, two aspects of the same thing. Its surface flows very easily under strong pressure, as for instance that which it meets in the jaws of stone crushers, but in this very flow it hardens greatly and rapidly, so that a relatively hard wearing surface forms, integrally united with a ductile back. Moreover, this surface is self-renewing, for as fast as a given layer is worn away, the next, as it thus becomes the surface, simultaneously hardens under the deformation which it at once undergoes.

The rapidity of the martensitization may explain not only the very small elastic ratio[3] but also the absence of necking during tensile test, with its

[1] *Op. cit.* Figs. 35 and 57, Plate 57.

[2] It is true that the Brinell hardness of low-carbon steel can be nearly doubled in wire-drawing, as shown in Table 25A. Hence I am hardly prepared to say that the greatest possible hardening by deformation is greater for austenite than for other metals. It is sufficient for our present purpose that the hardening caused by a given degree of deformation is far greater for austenite than for other metals, as is shown not only by the hack-saw experience cited above, but also by the rapidity with which manganese steel hardens in wire-drawing, changing with great rapidity to a hardness which is great compared with the initial. The mechanical support of the die increases greatly the deformation which can be endured without rupture, and thereby increases the possible hardening, for this increases with the deformation and hence with the ductility, as shown in Table 25A.

[3] Thus Professor James S. Macgregor of Columbia University finds in this alloy a proportionality limit of only 28,250 lb. per square inch against a tensile strength of 97,570 lb. per square inch (private communication, Aug. 28, 1913). In another case a proportionality limit of only 66,000 lb. per square inch, accompanied a tensile strength of 142,620 lb. and an elongation of 45 per cent. in 8 in. accompanied a contraction of area of only 36 per cent. (W. H. Stanger, Broadway Testing Works, June 21, 1893). The proportionality limit is that of the unmartensitized, the tensile strength that of the martensitized austenite.

consequence that the percentage of contraction of area may be much less than that of elongation. The necking which starts at the weakest section quickly arrests itself by generating, through the deformation which it represents, a martensitization so rapid as to make this section stronger than the rest.

Again Hadfield found that though in a tensile test of manganese steel the prick-punch marks tore so that the daylight could be seen through the test piece, yet this did not precipitate rupture, presumably because of this rapid strengthening of the metal in these regions of the greatest deformation.[1]

644. The Increase of the Apparent Hardness Caused by Varying the Brinell Test.—So, too, retarding the Brinell test may increase the hardness from 170 to 202.[2] The deformation starts a hardening process, which, as might be inferred from Muir's classical work,[3] and from Chapter 25, completes itself rather slowly. Hence retarding the Brinell test gives the step in this hardening process, started by each fraction of the deformation caused by the several successive increments of load, time to continue and thus to decrease the indentation which the later successive fractions of the load will cause, and thus in turn to increase the hardness number as found after the whole load has been applied.

My results, condensed in Table 24 here, show that the apparent hardness of manganese steel increases by nearly 20 per cent. on thus retarding the testing, whereas that of pearlite, the component of any hypo-eutectoid steel with which manganese steel would have to compete for its special uses, does not increase at all. On the other hand, the nearly pure ferrite of which the A.R.M. steel consists increased in hardness by no less than 37 per cent. Even after this increase it remains, of course, extremely soft.

TABLE 24.—INCREASE OF THE BRINELL HARDNESS CAUSED BY RETARDING THE BRINELL TEST

Material	Brinell hardness		Percentage increase caused by the retardation
	Under the usual direct test	Under a test greatly retarded	
Hadfield's manganese steel quenched from 1,100°. Steel No. 60.	170	202	19
A. R. M. Steel No. 0, carbon 0.01 per cent.	73	99	37
Eutectoid steel, No. 14, carbon 0.92 per cent., air cooled from 900°.	259	259	0

This failure of pearlite to gain hardness on retarding the Brinell test tallies with the observation that undivorced pearlite does not harden greatly in the tensile test as ferrite does. My experiments showed that the hardness

[1] The Author, "Manganese Steel," *Journ. Franklin Inst.*, Feb. 20, 1892, p. 6 of reprint. Compare Hadfield, *Proc. Inst. Civil Engin.*, 1887–1888, vol. 93, Part 3, p. 32 of reprint.

[2] Howe and Levy, *Trans. Amer. Inst. Min. Eng.*, 1915, to appear. In the retarded test a load of 1,600 kg. was applied at 10 a.m. and increased by 200 kg. every 2 hours till 6 p.m. when it reached 2,400 kg. At 11 the next morning and every hour thereafter the load was increased by 100 kg. till 4 p.m., when it reached the usual 3,000 kg. It was then left for the usual time, 30 to 60 seconds.

[3] *Phil. Trans. Royal Soc.*, 1899, A, vol. 193, p. 1, and especially p. 13 of reprint.

TABLE 25.—THE INCREASE OF HARDNESS DURING THE TENSILE TEST INCREASES WITH THE DUCTILITY

No.	Carbon content	Constituents	Thermal treatment	Brinell hardness: In the undeformed part	Brinell hardness: In the deformed part close to the grips	Brinell hardness: In the part close to the fracture	Increment per cent. of undeformed
1	0.01	Ferrite (A.R.M. steel) No. 0.	.....	77	140	159	106
2[b]	0.11	Ferrite + a little pearlite (sorbite) No. 15	.....	132[a]	169	195	47
3[c]	0.92	Pearlite (eutectoid steel) No. 14	Quenched from 775° and reheated to 600°	212 and 223[a]	.....	212	.....
			Quenched from 775° and reheated to 740°	163[a]		179	10
4[d]	0.92	Pearlite (eutectoid steel) No. 14	Furnace cooled from 775°	179[a]	.....	196	9
5[e]		Manganese steel Mn 12.8 per cent., No. 60	.....	217[a]	.....	340	57
6[f]		Manganese steel Mn 12.8 per cent., No. 60	.....	223	.....	540	145

[a] This result was determined in the least deformed part of the specimen, near the outer end of the grips, but there may have been slight deformation here.

[b] The Author, *Proc. Amer. Soc. Test. Materials*, 1914, vol. 14, p. 7.

[c] Cooled slowly from 1,100°, heated to 775°, quenched, and then heated for 3 hours to the temperature, 600° or 740°, as indicated in the table, and furnace cooled. The Author.

[d] Cooled slowly from 1,100°, reheated to 775° for 1 hour, and furnace cooled. The Author.

[e] Howe and Levy, *Bull. Amer. Inst. Min. Eng.*, No. 99, March, 1915, p. 591.

[f] Hadfield and Hopkinson, *Trans. Amer. Inst. Min. Eng.*, 1914, vol. 50, p. 486; and *Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, pp. 112 and 124.

TABLE 25A.—INCREASE OF HARDNESS CAUSED BY WIRE DRAWING AND COLD WORKING

No.	Authority and method	Chemical composition, per cent. C	Si	Mn	P	S	Shore hardness: Initial, A	Shore hardness: Cold worked, E	Shore hardness: Ratio, $\frac{E}{A}$	Brinell hardness: Initial, A	Brinell hardness: Cold worked, E	Brinell hardness: Ratio, $\frac{E}{A}$	Reduction of area in wire-drawing, per cent.
1	Goerens, wire-drawing....	0.11	0.003	0.56	0.044	0.044	24.0	46	1.92	118	201	1.70	48.3
2	" " "	0.12	0.148	0.56	0.016	0.017	26.0	57	2.19	135	265	1.96	70.4
3	" " "	0.27	0.167	0.43	0.023	0.015	22.0	58	2.64	130	249	1.91	74.2
4	" " "	0.62	0.223	0.65	0.045	0.017	33.0	59	1.79	185	326	1.76	44.3
5	" " "	0.98	0.158	0.43	0.025	0.015	59.0	62(52)	1.05	285	(361)	1.27	22.6(32.8)
6	Maurer, "cold worked iron."	....	.....	....	.....	.....	15.5	25	1.61	....	.....	.....	.....

Notes to Table 25A.—1 to 5, Goerens, *Carnegie Schol. Mem., Iron and Steel Inst.*, 1911, vol. 3, p. 410. 6, Maurer, *Rev. de Metallurgie, Mem.*, 1908, vol. 5, Fig. 12, p. 723; *Metallurgie*, 1909, vol. 6, Fig. 60, after p. 48.

of a broken tensile test piece, close to the fracture, was about twice that of the undeformed part in the case of the A.R.M. steel, which is nearly pure ferrite, and 47 per cent. harder than the undeformed part in the case of steel of 0.11 per cent. of carbon,[1] but that the increase was very slight in the case of steel of 0.92 per cent. of carbon, when its pearlite was undivorced, as shown in Table 25. This table shows that the increase of hardness in the tensile test increases in general with the ductility, that is, with the deformation. In like manner Table 25A shows that the hardening increases in wire-drawing with the actual deformation.

The foregoing results understate the hardening effect of the Brinell deformation on austenite, because the apparent hardness of this substance is exaggerated by the deformation incident to the test itself even when unretarded. This is shown by the progressive decrease which I find in the apparent hardness of previously undeformed austenite, with the load used in this test. Thus whereas lessening the load on the ball from 3,000 kg. to 500 kg. lessened the apparent hardness of the ferrite of the A.R.M. steel from 83 to 79, or by only 4 Brinell numbers, it lessened that of austenitic 20 per cent. nickel steel from 132 to 109 or by 23 numbers,[2] and that of Hadfield's manganese steel from 183 to 159 or by 24 numbers, using the normal 10 mm. ball, and from 215 to 182 or by 33 numbers when using a 6.35 mm. ball. The hardness of Hadfield's steel unincreased by deformation, as found by exterpolating the curve to a load of zero, should not exceed 125. These results are shown in Fig. 106A. The steepness of the left-hand end of these curves verifies my inference (§642) that a very great hardening is caused in austenite by the first stages of deformation.

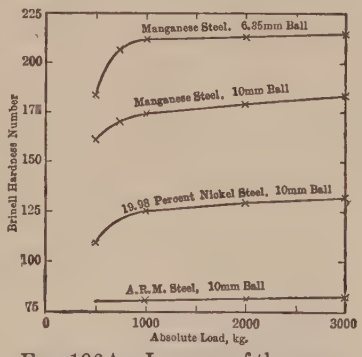

FIG. 106A.—Increase of the apparent Brinell hardness of austenite with increasing deformation.[3]

645. The Apparent Persistency of the Manganese Steel Lines.—When manganese steel which has been deformed is repolished and etched, bands appear which recall those formed by the deformation on a previously polished surface, much as the X bands, recalling slip bands, appear on repolishing and etching ferrite (§624), with the difference that these bands in manganese steel develop far more easily on etching than the X bands of ferrite. That such bands thus develop indicates that the metal, along the planes of slip throughout the mass, has undergone some change which causes it to react with the etching reagent differently from the undisturbed remainder of the metal. This inference is strengthened by Rosenhain's proof that the lines which

[1] The Author, *Proc. Amer. Soc. Testing Materials*, 1914, vol. 14, p. 28.

[2] The Unwin indentation tests of Carpenter, Hadfield, and Longmuir fail to disclose this hardening effect of deformation (*Seventh Report Alloys Res. Comm., Proc. Inst. Mech. Eng.*, 1905, pp. 892–3).

[3] The hardness, H, is calculated from Brinell's formula, $H = \frac{2W}{\pi D(D-\sqrt{D^2-d^2})}$. Where W = load in kg., D = diameter of ball in mm., and d = diameter of impression in mm.

etching develops in a similar form of austenite, 20 per cent. nickel steel, reproduce closely the pattern which the slip bands have formed in the same field.[1] The etching bands, in general like these, in specimens of this steel which apparently have not been deformed materially, suggest that congenital variations may thus simulate the effects of deformation.

This reminds us that Strauss[2] finds that very low-carbon steel which has been nitrogenized by heating in ammonia, when deformed even slightly and unintentionally and then polished and etched, shows martensitic markings, which therefore should represent a detectable change induced along the slipping planes by the deformation.

The bands developed in certain parts of some specimens of manganese steel by etching, without known deformation, may be much more prominent than those formed on a polished surface by moderate deformation. Note how very striking those of Figs. F and H of Plate 39 are. We shall see that these probably do not correspond strictly to those surface bands (§656).

646. The Apparent Persistency of the Manganese Steel Lines on Heating.—We have seen that the persistency of the X bands of ferrite on heating is limited, though it increases with the severity of the deformation which causes them. Thus Rosenhain finds that no slip-band traces of gentle deformation appear after exposure to 100°, and my X bands formed by punching, though they are still rather prominent after heating to 600°, disappear almost wholly after heating to 800° (§635). But the persistency of these manganese steel bands is far greater, for they appear even after a stay of 1 hour at 1,000°, followed by quenching, polishing, and etching. It was not the quenching that caused them in this case, because they did not appear after a like quenching not preceded by deformation. Moreover, heating may change their appearance greatly, without lessening their prominence. Note that in Fig. D of Plate 38, representing this steel heated to 575° after deformation, though as regards their direction the bands recall those of the etched but unheated specimens of Figs. A and B of Plate 39, yet they differ from those of the unheated specimens in being discontinuous, and in suggesting balling up.

The temperatures after heating to which these lines reappeared on repolishing and re-etching in the experiments of Howe and Levy are 400°, 550°, 575°, 600°, 800°, and 1,000°.

647. The sluggishness of this alloy, as regards the change from austenite into alpha iron, should be before us in interpreting the effect of heating on its deformation lines. Though at all temperatures above 750° it passes quickly into the nonmagnetic state of austenite, the reverse change into magnetic

[1] "Deformation and Fracture in Iron and Steel," *Journ. Iron and Steel Inst.*, 1906, II, vol. 70, pp. 212–213, and Figs. 12 and 13, Plate XLIII. The material is the Carpenter, Hadfield and Longmuir 19.91 per cent. nickel steel already referred to in §642. The specimen in which these slip bands had been developed was next etched deeply, and polished so as to remove the slip bands without effacing the deep etching figures, both operations being carried out while the specimen remained on the stage of the microscope with the original field constantly in view. The etching gradually developed lines exactly reproducing many of the slip bands, as the developer brings out a photographic image, and widened them out as is usual under all such conditions.

[2] *Stahl und Eisen*, 1914, vol. 34, p. 1814.

alpha iron is extremely slow, reminding us that in carbon steel also the change from ferrite to austenite on rising past $Ac1$ is incomparably more rapid than the reverse change on cooling past $Ar1$. Thus though the degree of magnetism which manganese steel can acquire by indefinitely long heating at 600° is estimated at more than 66 per cent. of that of pure iron, it reached only 27 per cent. in 10 hours at 600°, increasing by only 6 per cent. in the next 6 hours and by 1 per cent. in the following 9 hours.[1]

648. Causes of the Formation of the Etching Lines.—Slip having once altered the structure along the contact or slip planes, the persistent reappearance of the scars of that change along the slip contacts on repolishing and etching, either at once or after various intermediate treatments, may be caused by any one or more of several agencies, such as the amorphous iron generated by the slip; mechanical or annealing twins along those contacts; the contrast between austenite and the crystalline alpha iron resulting from the recrystallization of that amorphous iron; a greater proportion of alpha iron to austenite along those contacts than elsewhere; beta iron; precipitated cementite; and sonims. Let us glance at these.

649. Amorphous Iron.—Goerens' experiments[2] and mine described in §628 agree in showing that the effects of plastic deformation may persist even through a heating to 600°, and perhaps to a materially higher temperature. This persistence may reflect the persistence of unrecrystallized amorphous iron, and hence whether this is supposed to be alpha, beta, or gamma, its presence after reheating at least to 600° is a possible cause of the difference in etching tint along the slip contacts.

650. Recrystallized Iron.—Whether this amorphous iron is alpha, beta, or gamma, and to whichever of these forms it passes on returning from the amorphous to the crystalline state, its new crystals may well differ from those of the unaltered metal, whether in size or in orientation, and thus differ from them in etching tint. A temperature limit at which this feature of the scars must needs become effaced is not readily set.

651. Variation in Proportions of α, β, and γ Irons.—Moreover the amorphization and the later recrystallization may lead to an admixture of beta or alpha iron with the gamma along the contacts, an admixture which in this extremely sluggish material may not become effaced completely till the temperature passes above the transformation range, say to above 1,000°.

652. Twins may be set up along the contacts, mechanical ones at the time of slip and annealing ones on later heating. These are not likely to be effaced till the temperature is raised above $Ac1$.

The foregoing collectively may be regarded as closely allied to the martensitization which occurs in the hardening of carbon steel by rapid cooling. In

[1] Hadfield and Hopkinson, *Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, p. 109, and *Trans. Amer. Inst. Min. Eng.*, 1914, vol. 5, p. 476. They induced the greatest magnetism at 520°, and found $Ac1$ thermally at about 700°, on heating to which temperature nearly all the magnetism was lost. See also an important discussion of this subject by Sauveur, *Trans. Amer. Inst. Min. Eng.*, 1914, vol. 50, p. 501, and Fig. 31, VI, p. 180 of this book.

[2] *Carnegie Memoirs, Iron and Steel Inst.*, 1911, vol. III, p. 379.

one case as in the other, I refer the hardening proper chiefly to the formation of beta iron (§254).

653. Cementite.—Any transfer of the iron from the gamma to the alpha state caused by the deformation may perhaps itself precipitate cementite, which is nearly or quite insoluble in alpha iron, though abundantly soluble in gamma iron. Whether this cementite is or is not able at the room temperature to break away from its bond enough to affect the etching tint, it certainly can on any important heating. Hence as the temperature and time of heating increase, first the precipitated cementite coalesces into progressively thicker and more prominent masses, and second, till *Ac*1 is reached, the reversion of the gamma to alpha iron goes on slowly, and with it the precipitation of fresh lots of cementite. Any alpha iron formed in this heating toward *Ac*1 transforms on passing *Ac*1 into austenite, in which the redissolving of the precipitated cementite should begin at once, completing itself when *SE* of Fig. 23, p. 130, is reached.

Thus the quantity of precipitated cementite should increase progressively in heating till *Ac*1 is reached, and should decrease thence till the temperature attains the line *SE*, while through this whole period such precipitated cementite as is present should continue to assemble, thus exaggerating the prominence of the cementite masses till *Ac*1 is reached, and retarding their degradation during the rise from *Ac*1 to *SE*.

Each block of crystalline metal, being surrounded on all sides by metal which has been made amorphous by the disintegrating action of the slip along the slipping planes, would on heating act as an independent crystal, and expel to its outside all the cementite precipitated within it. This cementite should thus assemble along the very planes along which the slip has created amorphous iron, and because these planes are octahedral in austenite, it should form the familiar Widmanstättian figuring, so prominent in Figs. F of Plate 16 and A and D of Plate 38. The thickening of the manganese steel lines on reheating, shown in this last figure, naturally connects itself with this action.

654. Sonims.—This crowded migration of cementite would be likely to sweep along with it any inert matter, or sonims, such as the slag and manganese sulphide so often present in this steel. Though holding the temperature above *SE* should cause the reabsorption of the cementite, it should not affect these inert masses, which may thus become ineffaceable scars of the slip contacts.

655. Cementite not Martensite.—This heating cementite was naturally taken by Sauveur[1] for martensite, which it often simulates closely, both of them lying Widmanstättianwise along the octahedral planes like the amorphous iron. We have seen in §444 that the octahedral pattern of this cementite and of martensite are only two among many coördinate symptoms of a common cause, the octahedral structure of the austenite itself.

This white matter in manganese steel, shown in Figs. A and D of Plate 38,

[1] *Trans. Amer. Inst. Min. Eng.*, 1914, vol. 50, p. 508.

is proved to be cementite by its blackening with sodium picrate;[1] by its assembling, under favorable conditions, into a network closely like that of the pro-eutectoid cementite of hypo-eutectoid carbon steel; and by its being apparently identical with the corresponding cementite network which forms in this manganese steel, as shown in Figs. F of Plate 16 and A of Plate 38. Yet it may be accompanied by true martensite which it overshadows.

This cementite network forms the path of rupture in manganese steel, as in Fig. E of Plate 42, even more exclusively than in hyper-eutectoid carbon steel, and is probably extremely embrittling in both alloys.

656. A eutectiform pattern is shown in Figs. G of Plate 38, in which it occurs together with a marked dendritic structure; F and I of Plate 39, covering these figures completely; and C and H of this plate, covering these figures partly. It covered the whole of the specimen of Fig. B of this plate at first. I interpret it as representing the relics of a striped and initially continuous etching-pellicle, probably related to the dendritic structure of the casting, and broken up to varying degrees by the removal of parts or the whole of certain of its stripes on rubbing the specimen after etching.

Castings of this manganiferous austenite, even after quenching from 1,100°, and those of niccoliferous austenite,[2] may, on deep etching with either nitric or picric acid, become completely covered with a dark pellicle, such as occupies most of the large dark grain in Fig. H of Plate 39. This pellicle in manganese steel is striped in large part very regularly, but its striping is not readily shown by photography because the very dull colors of the stripes act much alike on the photographic plate. Slight rubbing, even the touch of the towel for drying the etched specimen, may remove certain parts of certain stripes, exposing the brilliant metallic surface below as white. A few such narrow stripes are seen in the upper part of the dark grain in Fig. H of Plate 39 already referred to. If a remaining stripe of the pellicle has certain transverse patches broken away, this causes the ladder-rung effect seen about ½ in. above the bottom of Fig. F. The breaking off of three or more adjoining stripes causes the irregular white patches seen in Fig. F. The prominence of the gray dendrites in Fig. G of Plate 38 I refer to the more complete rubbing away of this pellicle where they reach the surface than in the interdendritic spaces. The condition shown in Fig. H of Plate 39 results from unintentional rubbing; that in Figs. C and B of that plate from intentionally pushing the removal of the pellicle nearly or quite to completion.

[1] The Author, *Trans. Faraday Soc.*, 1914, vol. 10, p. 265.

[2] Carpenter, Hadfield, and Longmuir, *Seventh Report Alloys Research Comm.*, p. 945, and Figs. 27, 28, and 30, Plate 53, Nov., 1905. Their alloy contained nickel 19.91, carbon 0.41, manganese 0.96, silicon 0.13, sulphur 0.02 and phosphorus 0.01. Professor William Campbell preceded me in detecting this structure in manganese steel. To him I owe the method of deep etching with picric acid as a means of developing it. A micrograph by Hadfield and Hopkinson, their No. 3, shows this pattern in manganese steel of "the usual water-toughened form with fine-grained polygonal structure," "usual" evidently referring to the water-toughened state and not to this pattern (*Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, p. 116, and Plate IX). None of these investigators explains the cause of this pattern.

This striped pattern may be replaced by others, for instance the crackled one of Fig. I.

657. Is the Eutectiform Pattern Related to the Structure of the Underlying Metal?—I find three indications that it is. First, the great vertical gray dendritic column of Fig. G of Plate 38 is evidently related structurally to the rectangular pattern formed by the oblique dendrites which lie on either side of it in that same grain, and run about N-25°-W and N-65°-E, for its figurings follow these same directions, and indeed are often prolonged as parts of this oblique pattern. In this column itself there are gray and black stripes which connect themselves with the eutectiform pattern, and also are parallel to the oblique dendritic pattern. Moreover, the eutectiform pattern itself within the dendritic rectangles has two main directions, of which one, followed in the greater part of each rectangle, is parallel to the vertical column, and the other, followed in the immediate neighborhood of the oblique branches themselves, is parallel to them. In many of these rectangular spaces the eutectiform stripes can be seen plainly swinging round from the vertical to N-65°-E as they approach the great oblique dendrites.

In the grain which occupies the narrow vertical strip from $\frac{3}{16}$ to $\frac{3}{8}$ in. wide at the right-hand side of this micrograph, the relation of the eutectiform pattern to the black dendritic bat's wings is even more clear.

Second, the general shape of the individual areas in which the eutectiform pattern follows an approximately constant direction, is very often that of the polyhedral grains of austenite, and moreover these grains are often marked out by a distinct black boundary, as true grains so often are. This is seen in Figs. H and I of Plate 39.

Third, the main directions followed by the pattern in Fig. F of Plate 39 are those which the structure of the metal itself would follow. The bottom and left-hand side of this figure represent two sides of the casting itself, from which columnar crystals normal to the two cooling surfaces might well shoot, with diagonal bandings filling the space between quite as they do in this pattern.

The honeycomb structure shown in Fig. I is less clearly connected with the solidificational structure of the metal.

The sharp change of direction between areas 1 and 2 of Fig. H, suggesting twinning, and the shape of many of these areas, recalling those of the pearlite colonies of Figs. A and C of Plate 1, further dispose us to regard the pattern as related to the structure of the underlying metal.

658. Does this Structure Represent a True Eutectic?—Certain stripes of the etching pellicle certainly seem to adhere rather more strongly than those with which they alternate, for otherwise the regularity of this pattern would be hard to understand. Do the more and the less firmly adhering stripes overlie the alternate bands of a eutectic? This question cannot be answered without further study. But the fact that in Figs. B and C the martensitic structure, which underlay the pellicle, seems much more regular than the eutectiform pattern itself usually does, may perhaps mean that the two con-

stituents of the eutectic, though they differ in chemical composition, are isomorphous and identical crystallographically, so that they are affected alike by the quenching stress to which the martensitic structure may be referred, and so that this structure passes across them as if they were one.

659. Eutectic or Eutectoid?—As between a eutectic and a eutectoid, this structure is more likely to represent the former than the latter, because it is present in specimens quenched from 1,100°, a temperature at which the constituents of any eutectoid would probably merge into each other, as those of pearlite merge and form austenite when heated above *Ac*1. This inference can be tested by noting whether this structure occurs after much forging, for this should distort and might efface the eutectic structure.[1]

Because the niccoliferous austenite in which it occurs contains little but nickel and iron, the eutectic, if it is one, is likely to be composed of different nickel-iron alloys in that case and of different manganese iron ones in my case.

660. Twinning Suggested by a Change in the Relative Prominence of Two Conjugate Sets of Manganese Steel Lines.—Both among the surface bands and the etching bands there are often two conjugate sets, certain individuals of which stop or lose prominence on intersecting a line of the conjugate set, or lose prominence on entering a certain area. Thus in Fig. D of Plate 39 two S-60°-W lines running from *c* to *d* at first look as if they stopped at *d* and changed direction, whereas in fact they continue their direction down beyond *d*, but become fainter at the left of *d*. So the lines running S-60°-W from *f* at first seem to stop at *e* and change direction, whereas in fact they extend beyond *e* but lose prominence. Again, at the imaginary line *ab* a great number of the lines running S-60°-W seem to twin into corresponding lines running N-45°-W, as if *ab* were the boundary of a twinned area. Yet if we extend *ab* a little higher up we see that these two sets of lines in fact continue beyond their intersection. Again, in the etched Fig. B of Plate 39, though each of the three prominent sets of lines, running N-80°-E, N-40°-E, and N-30°-W respectively, is strongly marked only in certain areas, yet each set can be traced in other areas in which it is less prominent.

Further study is needed to determine whether these are really twins. We saw in §558 that this reversal of the relative prominence of two sets of slip bands is a frequent characteristic of twinned zones.

661. Discrimination between the Various Forms of Manganese Steel Lines.—Our first impulse is to regard all these lines as one and the same thing, because of their family resemblance. But we have now seen reason to infer that the etching bands found after heating, Figs. F of Plate 16, A and D of Plate 38, and F of Plate 42 represent coalesced cementite, and that the

[1] Carpenter, Hadfield, and Longmuir do not refer to the presence of this structure in any of their forged or otherwise much deformed specimens. It is suggested in one of their bent specimens (their Fig. 50), but the deformation of this need not have been enough to affect this structure markedly. Howe and Levy show a related structure close to the fracture of a tensile test piece ("Are the Deformation Lines in Manganese Steel Twins or Slip Bands?," *Trans. Amer. Inst. Min. Eng.*, 1915, to appear, Plate 3, Fig. 12), but in neither case is it clear that the deformation would suffice to break up a eutectic structure composed of two alloys of closely like properties.

zebra bands of Figs. F and H of Plate 39 represent a eutectic and hence are not deformation lines. Moreover, we found difficulty in §627 in determining the relation between the slip bands of ferrite and its X bands found on etching. Hence, though all these markings, except the eutectiform ones, are likely to be scars of one kind or another resulting from the deformation along certain octahedral slip planes, yet they need not represent the same feature of that deformation. The surface bands may represent slip which is accompanied or followed by twinning, and the scars left by this twinning and those which result from the slip proper will naturally have the same octahedral pattern. Let us therefore ask with some caution what the surface and the etching bands represent.

662. The Surface Bands are Slip Bands.—These lines might be either slip bands or Neumann bands, that is, mechanical twins. We may infer that they are the equivalent of what, in other metals, are called slip bands for four reasons, A to D, which I will now give. These reasons are equally valid whether we regard slip bands as representing strictly geometrical or "vectorial" translation, or as representing also fluid motion in greater or less degree.

663. (A) **The Course of Their Development.**—With progressively increasing deformation, the first appearance and the development of these surface bands are like those of the slip bands in other metals, as is shown by applying to iron, copper, and lead the truncated wedge experiment outlined in §413. This resemblance is shown by a comparison of Figs. B, E, H, and K of Plate 38, which illustrate the development of these bands in manganese steel, with Figs. C, F, I, and L of the same plate, which illustrate the development of the slip bands in copper.

Save that copper roughens more than manganese steel, one metal might often be taken for the other, so closely do the manganese steel bands and the copper slip bands correspond in their development. For instance, the order of regularity of the two sets of bands in Fig. H and their behavior at their intersections are closely like those of the corresponding sets in Fig. I.

These bands in copper are accepted as slip bands. Those in manganese steel start and develop as closely like those in copper as could be expected in two different metals. The differences are rather of the order which we expect between identical phenomena in different substances, such as the difference between the waves in oil and those in water and ether, than such as we find between radically different phenomena.

Moreover, these surface bands in manganese steel correspond closely to the slip bands of other forms of austenite, such as invar, in shape and general appearance.

One gets the impression that the development of the individual surface bands in manganese steel is progressive, so that a given band, starting as a light and short mark as in Fig. K of Plate 38, lengthens and gains prominence. Should this prove true, it would agree better with our conception of the nature of slip than with that of twinning. Slip once started along a

given plane, might continue there, and a given instantaneous throw of slip might be great or it might be small. But the throw of a twin of given width should, in a given area, be fixed and incapable of increase. Twinning is a self-completing action, arresting itself at a certain predetermined end, whereas the distance covered by slip is indeterminate.

664. (B) **Their Relation to the Twinned Zones.**—The surface bands and the twinned zones of Fig. B of Plate 29 bear to each other the same relation as regards their order of magnitude that slip bands bear to twinned zones not only in copper, silver, and lead, but also in 25 and 36 per cent. nickel steel or invar, two forms of austenite closely comparable with manganese steel. Indeed, these lines and the twinned zones in Fig. B of Plate 29 might easily be taken for slip bands and twinned zones in copper. That the relation of the surface bands to the annealing twins in manganese steel is like that of the slip bands to those twins in invar is seen by comparing Figs. A, B, and C of Plate 29 with Figs. D, E, and F. Note, for instance, that in both Figs. A and E lines of the direction which is most prominent in the odd-numbered areas occur but with less prominence in the even-numbered area or areas, and *vice versa*.

665. (C) **Slip Bands to be Expected.**—Though slip bands can be developed abundantly without twinning, in the ductile metals in general, yet so far as we know any plastic deformation which causes twinning must also cause slip bands, which thus seem to represent the normal and easy mode of translation. Thus the quiescent deformation which creates the slip bands so abundantly in alpha iron gives no suggestion of Neumann bands, whereas the shock which develops the Neumann bands always develops slip bands also, and nearly always abundantly. Even in coarsely crystalline steel in which the Neumann bands form very readily, they are always accompanied by some slip bands. Hence when, as in manganese steel, only a single set of lines usually results from even great plastic deformation, analogy indicates that these lines are more likely to be slip bands than twins.

It is in conformity with these ideas that a brittle metal like bismuth, which is incapable of much plastic deformation, forms Neumann bands very abundantly. And it is not out of harmony with them that, in the brittle silicon steel, certain areas which have Neumann bands are nearly or quite free from slip bands, because we have as yet no evidence that appreciable plastic deformation has occurred here.

666. (D) **Hardening Points to Slip.**—The amorphization and martensitization (§668) along the planes of which these manganese steel bands are the outcrops, implied by the rapid hardening, tally with the belief that slip has occurred along them but not with the belief that only twinning has, as is explained in §619; for, first, slip is naturally believed to cause such amorphization, whereas twinning is not, because it implies only rotation into a new orientation without crushing any crystalline matter, and, second, like reasons point to slip as the cause of martensitization.

667. (E) **The Great Deformation Suggests Slip.**—Another consideration, which at first seems to argue that these lines must be slip bands, on examina-

tion loses some of its force. That consideration is that in other metals we refer the deformation primarily to slip bands, first because the twinning does not occur during the deformation, and second because for other reasons it seems incompetent to explain the deformation. This might suggest at first that if the manganese steel lines are all of one kind, they must be slip bands in order to explain the deformation. These reasons apply with less force here, first because the manganese steel lines form during the deformation itself, and second because of their abundance. Yet, even so, the very great deformation which this material can endure without rupture refers itself more readily to slip than to twinning, because the ratio of the extent of the possible motion to the thickness of the altered layer is so much greater in slip than in twinning.

668. (F) The Apparent Persistency of These Surface Bands is Compatible with Their being Slip Bands.—By their apparent persistency I mean the fact that, on repolishing and etching, bands appear recalling the surface bands closely. That this resemblance is compatible with their being slip bands follows from what has already been said, for it means only that etching develops a pattern which resembles that of the slip bands in a general way. But this is wholly inconclusive, for deformation by slip, if it causes true slip bands, may also lead in any of the ways indicated in §§648 to 654, to the formation of etching scars or traces following the same patterns, whether there is or is not heating between deformation and etching. The resemblance, whether rough or exact, means only that the marked octahedral structure of the austenite is followed by the slip and also by twinning if that accompanies or follows slip, and by the various immediate causes of scars, the ulterior cause of all of which, as of the slip bands, is deformation along one and the same set of octahedral planes.

The reason why slip in ferrite often causes no pattern on etching is not clear. But the differences that ferrite may perhaps twin under quiescent deformation far less readily than austenite, and that on heating it neither reverts to another allotropic form nor generates cementite as this specific high-carbon austenite does, are enough to explain why these etching scars occur always in this austenite but not always in ferrite. This supposed more ready quiescent-mechanical twinning of austenite could in and by itself explain the persistence of its surface bands, on the hypothesis that they represent slip accompanied by mechanical twinning along the slip planes. The most attractive explanation is that the slip is accompanied by martensitization along the slip planes in this alloy. This would explain why etching invar, 36 per cent. nickel steel, after repolishing it, does not develop bands simulating the slip bands, because this alloy has so great an excess of nickel, and hence is such a stable form of austenite, that it ought not to martensitize under deformation. (Note 2, p. 182 and §642.)

669. The nature of the unheated etching bands in manganese steel, that is, of the bands which etching develops when it is not preceded by heating, cannot be inferred with complete confidence from any of this evidence, for

the reasons just given. The straightness and regularity of these bands in Fig. B of Plate 39, and their family resemblance to the Neumann bands in ferrite of Figs. C and D of Plate 32, in zinc, Fig. A of Plate 35, and in bismuth, Fig. D of Plate 41, are not conclusive, because in these same respects these etching bands of manganese steel resemble equally the surface bands. The indication which led Osmond and Cartaud[1] to interpret such etching bands as Neumann bands, that they seem to be a reappearance of the surface bands, is equivocal for the reasons given in the last section. Indeed, the curliness of the etching bands in Fig. A of Plate 39, taken with the fact that in another field in the same specimen like bands are as straight as they are in Fig. B of this plate, tends to connect them rather with the surface bands of manganese steel, which may curl, than with the Neumann bands of other metals. On the other hand, initially straight Neumann bands might well be curled by later deformation in so ductile a material. Though the martensitization, with reversion to the hard beta state, implied by the rapid hardening of this steel on deformation, suffices to explain these bands and is to be referred to slip, it may be accompanied by twinning. We can learn whether they are twins, as their uniform width suggests by noticing whether they zigzag later formed slip bands.

670. The nature of the heated etching bands, that is, of those found by etching specimens which have been heated after deformation. Markings like those of Figs. A and D of Plate 38, found after heating to moderate temperatures such as 575°, have been shown to consist of cementite (§655). Further study of those found after heating to 1,000° and quenching is needed. It is possible that even these are of cementite, for 1,000° may be below *SE*, at which the reabsorption of cementite is due. Whatever their true meaning, they remind us of the persistency of this Widmanstättian figuring under other conditions,[2] and of that of the initial dendritic casting or ingot structure,[3] both of which persist through treatments which at first would be expected to efface them. In some cases this persistency may represent simply a defiance of surface tension. In others it may be referred to the co-concentration of minute quantities of impurities, such as phosphorus, which diffuse very slowly, or of sonims which do not diffuse. In this present case, the persistence of the manganese steel lines through a long sojourn at 1,000° and quenching may be due to the latter cause, lines of impurities once swept along by the migrating cementite into the slip plane contacts where the mar-

[1] *Journ. Iron and Steel Inst.*, 1906, No. III, vol. 71, pp. 468–9.

[2] Howe and Levy showed that this structure has extraordinary stability in an 0.40 carbon steel, suffering no important breakdown on staying either for 137 hours at 650° or for 21½ hours at between 725° and 740°, though in this latter case the very strong cementite network structure of a 1.14 carbon steel beside the Widmanstättian steel broke down completely. "Notes on Divorcing Annealing and Other Features of Structural Coalescence in Iron and Steel," *Proc. Cleveland Inst. Engineers*, July, 1914, pp. 234–243; and Stead, *idem*, p. 252. The Widmanstättian structure of meteorites has evidently persisted through the almost infinitely long opportunity for its break-up during their secular cooling (§345).

[3] Engineer N. T. Belaiew, *Revue de Metallurgie, Mem.*, 1912, vol. 9, p. 647. Cp. §§783, 796.

tensitization occurred, remaining undiffused long after the cementite which dragged them there has on heating diffused away and been reabsorbed, after the amorphous iron has reverted to the crystalline state and after the martensite formed by the deformation has changed back to gamma iron.

Or, more generally, these lines may simply represent the incompleteness of the effacement of the scars formed along the slip planes. Because this effacement implies the close identification, both chemical and structural, of the altered and unaltered material, it may well be tardy. A structural difference caused by the deformation might well persist through varied structural changes, quite as the initial lead of a runner might persist through numberless doublings of the path to which he and his chaser are closely confined.

Thus the surface bands are slip bands; the unheated etching bands which deformation causes are shown by the accompanying hardening to represent in part martensitization with the formation of beta iron caused by slip, though, as their uniform width suggests, they may represent twinning also along the slip planes. Heating precipitates cementite along these same planes, so that the heated etching bands represent this cementite, either alone as in undeformed specimens, or in addition to what the unheated etching bands represent in deformed ones.

CHAPTER 27

IS THERE FLUID AS WELL AS CRYSTALLINE MOTION IN METALS?

671. Introduction.—The block movement by means of grain uplift, slip, and twinning which has been considered in Chapters 16, 18, 22 and 23 might conceivably be accompanied by fluid movement, which need not in itself cause any of these three varieties of block movement.

Let us first sum up the evidence of crystalline motion given by the twins, the Neumann bands, the straight and crystallographic parts of the slip bands, and the etching pits, and next the evidence of fluid motion given by the curved and irregular parts of the slip bands, their profiles, their optical properties, and by the other rounded foldings. Let us then consider how this conflicting evidence can best be reconciled.

672. Evidences of Crystalline Motion.—(A) *The twinned areas* show that they have been formed by crystalline motion by the straightness of their boundaries, and by the identity in all the odd-numbered ones, and the corresponding identity in all the even-numbered ones, first of the direction of the slip bands, second of the shape of the etching figures, and third of the etching tint under oblique light (§§558, 561 and 562). Twinning is more persuasive than slip as evidence that the motion is geometrical and hence crystalline, because it implies that the existing crystalline conditions govern not only the direction but also the extent of the motion, arresting it when it has reached a point which they predetermine.

(B) The straightness of the *Neumann bands* (§571), their following all the directions of the planes of the 211 trapezohedron and perhaps of the cube, but no others (§§589 to 591), and the constant relation between their internal orientation and that of the enclosing metal (§601) give strong evidence that the motion which accompanies their formation is preponderatingly if not wholly crystalline.

(C) The straightness of certain *slip bands* (§§419 and 441), the accuracy with which they follow definite crystallographic directions (§§440 and 441), and the occasional sharpness of their treads and risers (§§418 and 466) tend strongly to show that they represent crystalline motion.

(D) The many cases of regularity of the *etching pits* (§373), even after great deformation, show that the orientation of the metal is geometrical and hence crystalline after that deformation. Because such regularity can hardly have arisen after the deformation, it argues that the deformation itself has been vectorial and hence crystalline, at least in very large part. But this test is not very sensitive because the etching pits after all are only roughly rectangular, so that a very appreciable degree of unvectorial or fluid motion may have occurred.

(E) *The uniformity of etching tint* throughout any given grain, and the sharp change from grain to grain, even after great deformation, because they imply that the orientation is constant throughout each grain, in like manner argue that the deformation has been preponderatingly vectorial, because, had it been in large part fluid, it could hardly have avoided disturbing the initial uniformity of the orientation in each grain. This test, like that under D, is not very sensitive.

Thus we have evidence that the motion which generates the annealing twins and the Neumann bands is preponderatingly crystalline. And though the evidence under C applies only to certain slip bands of early deformation, yet under D and E, because the motion which preceded these observations must have been in very large part that which causes slip bands after great deformation, we have evidence that the motion which generates these also is preponderatingly crystalline.

In addition to this strong evidence there is some which points, though less strongly, to crystalline motion. Let us glance at this.

(F) The close localizing of the early stages of the deformation of lead about its slip bands and the close reproduction by etching of the deformation figures set up on the polished surface of austenite (§§474, 645) are such as would occur were the motion strictly crystalline.

But this is equivocal, because a like localization, which sometimes appears to be very complete, occurs about the Lüders lines, though these do not represent crystallographic directions, recalling the less complete localization about the folds in varnish and cloth.

(G) The weakening effect of grain coarsening (§§489 to 494), though it was predicted as a result of the crystalline nature of the deformation and hence tends to support the belief that there is crystalline motion, yet is so moderate that it is compatible with the belief that there is also much fluid motion. This is true also of the hardening, strengthening, and like effects of plastic deformation. They tend to show not only that there is crystalline motion, but more particularly that there is slip as distinguished from twinning. Yet they too might occur even if a large fraction of the motion were fluid.

(H) The evidence indicating the progressive strengthening of the metal about the grain boundaries as deformation progresses (§§407, 495 and 694) argues for the occurrence of slip, because that strengthening is a natural result of the propagation of slip across the grain boundaries to disregistering slip planes beyond.

673. Evidences of Fluid Motion.—Let us turn now to the indications of fluid motion. The many roundings of the surface which have been cited in §§410 and 411 suggest fluid motion strongly, and they are not accounted for by any visible signs of block movement. (See, for instance, Figs. E of Plate 24, F and G of Plate 25, D of Plate 27, A of Plate 31, and A and C of Plate 32.)

The straight and crystallographic parts of the slip bands may be accompanied, as we saw in §419, by parts rounded in plan even when they are

developed in the absence of important previous deformation, while after such deformation the direction of the slip bands in plan may be hopelessly irregular (§§419 and 463). Moreover, their profiles give suggestions of being rounded, both by their appearance in cross-microsections (§466) and by their optical behavior, in that they reflect vertically light which comes from a wide range of obliquity, in that those which do this form a large fraction of the total visible under vertical illumination, and in that they very often reappear on 180° rotation (§§467 to 470).

The irregularities and roundings of the silhouettes, quite like those of the slip bands of which they are assemblages and for the same reasons, suggest a large proportion of fluid motion (§§451 and 452).

Again, the irregularity of profile of the evidently crystalline Neumann bands, and especially the rounding of the re-entrant profile angles, make a like suggestion of fluid motion (§§579 to 580). (See Figs. A to D of Plate 36.)

These evidences of fluid motion are not indeed conclusive. Of the two essential parts of which they consist, one, the roundings and irregularities, is explicable as representing either assemblages of conjugate crystallographic straight lines so short as to integrate into apparently smooth curves or the effect of crystalline deformation along a new set of planes after the lines and bands which we find deformed have initially formed. The other, the optical behavior of the slip bands, is explicable as the result of the rounding of their profiles by smearings of the amorphous metal generated along their treads by the prior polishing, and along their risers by the slip itself as a sort of slickensides (§§467, 470 and 471), and perhaps as a result of rounded folds in the amorphous surface layer developed by the polishing. For that matter, the discrepancy between the optical behavior of slip bands and the hypothesis that they represent pure crystalline slip is suspected only, and this suspicion may be removed by more thorough study.

Up to this point we find on one hand a great mass of evidence, all of it equivocal, suggesting a large fraction of fluid motion, opposed by strong evidences of crystalline motion, and especially by the regularity of the etching pits and etching tint even after great deformation, indicating that the motion is preponderatingly vectorial. The reported greater irregularity of the etching pits after than before great deformation (§373) carries little weight: first, because it is lacking in detail; second, because even in strictly crystalline material the etching pits may be very irregular on random planes; and third, because, though a perfectly oriented mass might well yield irregular pits even on crystallographic planes, a disoriented mass cannot possibly yield regular ones.

Because the evidence of vectorial motion is so strong and that of fluid motion equivocal, we infer that in these cases in which the etching pits are regular even after extreme deformation, the motion has been preponderatingly vectorial.

674. A suggestion of fluid motion is given by the progressiveness of the change of etching tint of the narrow Neumann lamellæ of Fig. A of Plate 45,

in a bar which has been twisted about its axis after the development of these lamellæ. The suggestion here is that each of the crystal units which make up each lamella, in rotating about the axis of the bar, has also rotated slightly about its own axis, and thus has moved in part progressively unvectorially, that is fluidly. But I find quite as progressive a change in etching tint in a twin in a crystal of native copper, Fig. AA of Plate 30, which has undergone such slight deformation that fluidity of motion is hardly to be looked for, especially because the slip bands set up after incomparably greater deformation, and the twins which have persisted through that deformation, show no signs of fluid motion (Figs. D and F, Plate 45). This phenomenon needs further study.

675. Even this Evidence Shows that the Motion is Preponderatingly Crystalline.—Even this strong suggestion that there is an element of fluidity in plastic deformation is accompanied by evidence that the deformation is preponderatingly if not overwhelmingly crystalline, in the sharpness and perfect continuity of these attenuated Neumann lamellæ. Here we must remember that, because their creation does not affect the strength, and because they are symmetrical in orientation with the enclosing metal, their substance is probably in immediate contact with that metal, unparted by any amorphous or other sheath, and hence their boundaries are almost purely imaginary surfaces as far as the path of deformation is concerned, quite as the twinning line *AB* of Fig. 93, p. 402, strongly as it strikes the eye, is yet almost purely imaginary as regards the properties of the pavement. This is true at least for the purpose of my present contention that, were there any large element of fluid motion here, the twisting would set up eddies which would disregard the surfaces of these lamellæ almost completely and break them up into wavelets and vagueness. Because these lamellæ show no sign of such vagueness or fluid irregularity, but on the contrary remain perfectly sharp in spite of their narrowness, and because such deformation as occurs in them suggests sharp crystalline faulting, as at the left of *S*, Fig. A, I infer that even here the deformation is preponderatingly crystalline. It is indeed surprising that the etching has left the edges of these lamellæ so smooth. The retention of continuity by a lamella cannot be referred to any special strength, as of a cobweb floating in air or of a thread in porridge, and therefore this visible continuity testifies to the persistence, through the twisting, of the continuity of every imaginary and invisible stratum, and thereby suggests that the deformation is preponderatingly crystalline.

676. Additional evidence that deformation is preponderatingly crystalline is given by Figs. D and F of Plate 45, which show the transverse and longitudinal sections of a bar cut lengthwise from a crystal of native copper elongated by parallel growth, $\frac{3}{8}$ in. (9 mm.) long between shoulders, and twisted 360° about the axis of the crystal, which is the one shown in Fig. E of Plate 45. This bar was deformed so as to create the twinning tendency; annealed at 600°–650°, so as to cause twinning; twisted about its axis as just described; cut and polished longitudinally and transversely; and again

deformed so as to set up slip bands and disclose by means of them the present shape of the twins. That the twin of Fig. F of Plate 45, which was formed before the 360° rotation, is still straight after it certainly argues forcibly that the deformation in this rotation has been strictly crystalline. The straightness of the slip bands in that figure points in the same direction.

Very strong evidence is given by the geometrical relation between the three directions of slip bands in the longitudinal section, Fig. D of Plate 45. Two of these directions are prominent, and the third is indicated by means of the broken line *ab*. The angles between these three directions are 62°, 68°, and 50°. Their closeness to the theoretical angle of 60° is certainly striking in view of the difficulties of measurement and of making the plane of the microsection coincide with the crystalline axis of so minute and deformed a crystal-fragment as this, for it is only in an accurately crystallographic plane that the three angles should be accurately equal and 60°. It is indeed more by a hit than any good wit that this coincidence is as good as these observed angles imply. These angles measured in another field in this same section are 63°, 63°, and 53°.

The fact that, after this 360° twisting, the slip planes retain thus accurately their initial inclination to each other, indicating as it does that the twisting has not disturbed the geometrical crystalline relations between the component particles of the mass, certainly argues strongly that this plastic deformation has been at least preponderatingly crystalline. A like argument is given by the fact that the straightness of these slip bands after this twisting is of the same order as that which preceded the twisting, as shown in Fig. E of Plate 45 of this same specimen.

The inference that in these cases the deformation, even when pushed far, is preponderatingly crystalline, makes it easier to interpret the irregularities of the slip bands found in other cases as resulting from irregular stepping back and forth between conjugate sets of slip planes and like causes.

677. The Progressive Accumulation of Amorphous Metal during Deformation should Cause the Motion to Become Partly Fluid in Essence.— If, with Beilby (§§508 and 509) we conceive that, after the first fraction of slip, every crystalline block which has moved ever so little relatively to its neighbors is thinly sheathed with amorphous and temporarily mobile metal, and that each succeeding fraction of slip breaks these blocks into smaller and smaller strictly crystalline fragments each with its amorphous sheath, we find a reason why the initial regularity of the slip bands should be replaced by progressively increasing irregularity as deformation goes on.

For though, in the first stages of deformation, when the continuity of the crystalline masses is interrupted only at few and distant points by the little amorphous metal hitherto generated, slip tends strongly to follow along that same crystalline plane in which it starts, with the result that a given slip band remains straight from side to side of a given grain, yet after advanced deformation has set up innumerable sheaths of amorphous metal about immeasurably small residual crystalline blocklets, this tendency toward

straightness is much less, because of the immeasurable opportunities which slip has to turn from its initial direction to a conjugate one, the two jointly conforming more closely to the line of stress than either could alone. Such an opportunity comes when slip, on leaving any one crystalline blocklet, enters the enveloping amorphous sheath which parts that blocklet from the next.

This progressively increasing frequency of opportunities for changing direction may well lead to a progressively increasing deviation from the straightness which the slip bands in the early stage of deformation show, resulting in time in the hopeless irregularity of Figs. C and E of Plate 22.

In the same way may be explained the irregularity of the silhouettes after great deformation, for these silhouettes after all are composed of slip bands, and a cause which deforms individual slip bands may well be competent to deform to a comparable degree these assemblages of slip bands.

In its passage across any given amorphous sheath, the deformation would naturally be fluid; but it is probable that these sheaths are extremely thin, and form a very small part of the total mass even after great deformation, so that the part of the path of deformation which actually lies within them is a very small fraction of the whole. This leaves us with the conception that, after deformation has gone on to any considerable degree, its path consists of two distinct parts, a very small fluid part in the thin amorphous sheaths and a very large crystalline part composed of straight fragments of increasing shortness and following conjugate directions, and thus, together with the fluid fractions, integrating into lines which become more and more irregular as deformation progresses, and hence suggest fluidity increasingly. The degree of true fluidity suggested by the progressive change in etching tint in the Neumann lamellæ of Fig. A of Plate 45 is not too great to be given by means of these thin amorphous sheaths.

These considerations point to slip rather than twinning as the mechanism which causes the slip bands, for the reason so often given, that amorphization seems a more probable result of slip than of twinning (§619).

678. This principle may explain the progressiveness of the change in etching tint in twisted Neumann lamellæ, as in Fig. A of Plate 45, for though in the twisting before this etching the total deformation was relatively slight, it was cumulative as regards its effect on the rotation of the little crystal units, every fraction of it tending to rotate them in the same direction, whether this rotation is solely of those units between their amorphous sheaths as I have suggested, or whether there is an additional mode of fluid motion.[1]

[1] In asking why this fluid component of the motion which is shown so equivocally in other cases is so strongly suggested here, one might conceive that the progressive rotation of these units represents only the elastic stage of the deformation. To bend one of these lamellæ, if it could be isolated, about the axis of the bar in exactly this way would not strain it beyond its elastic limit, because they are only about $\frac{1}{3000}$ in. thick, like tissue paper, and yet it would rotate the crystal units in exactly this way. Clearly the motion during the elastic stage of such deformation is in effect unvectorial. But this conception is to be rejected because the fact that the Neumann lamella is integrally united with the rest of the bar reduces the elastic part of its deformation to an insignificant part of the twisting of the whole, a part too small to account readily for the progressive change of orientation.

In cold rolling and punching, on the other hand, though the total deformation is very great, it does not tend to tip the crystal units more in one direction than in another, so that the various fractions of whatever fluid motion occurs might well tend to cancel each other.

679. Why are the Etching Pits so Much More Regular than the Slip Bands after Great Deformation?—The case is more difficult than that of the uniform orientation of any given spheroid of a liquid crystal as shown in Fig. B of Plate 14, because this latter case does not imply the retention of the initial crystalline orientation throughout the mass while surface tension is in process of making a spheroid of it. Indeed the initial orientation may, during the spheroidizing be lost by those particles which are actually migrating, and be re-established as regards those pilgrims, after they have come to rest, by propagation from the central and undecrystallized mass.

But the viscosity of cold deformed iron is so great that we hardly admit the probability of such recrystallization by propagation. Hence we infer that, even after severe deformation has broken the mass up into immeasurably small crystalline blocklets, each with its amorphous sheath, the mass of the sheath is so extremely thin that it acts only as a lubricant, with the result that the tendency to close packing causes each blocklet to retain its orientation, as each one of a mass of greased cubes might retain its orientation during the deformation of that mass as a whole. The amorphous sheaths about these crystalline masses may indeed suffice to deflect the path of deformation along an enormously great number of pathlets along true even if immeasurably short conjugate crystalline slip planes, integrating into a crooked line; and after a slip band has formed its irregularity may be increased by faulting caused by one or more sets of later slips, each slip being unrecognizably minute; and yet the orientation of the crystalline fragments may remain but little disturbed, with the result that they etch into rectangular pits. The rough rectangularity of the etching pits is compatible with a considerable degree of change of orientation of certain particles. Under favorable conditions, as in the drawing out by cold rolling which preceded the formation of the regular etching pits of Fig. E of Plate 18, the disturbance may be slight, that in one stage cancelling that in another. Under other conditions, such as those of the twisting of the Neumann lamellæ of Fig. A of Plate 45, it may be greater because cumulative, each step in the rotation of a given crystal unit within its sheath being added to all the other steps in causing the resultant final rotation; or because the mass is so thin that the directive force exerted on any one crystal unit by its neighbors is slight. This last suggestion has some support from the fact that change of tint in the Neumann lamellæ of this figure is much greater than that of the enclosing metal along this same direction, as if the rotation of the crystal units of the lamellæ were exaggerated by some special influence.

Whatever the merits of this hypothesis, it is sufficiently wonderful that these crystalline blocks can within wide limits follow the deformation of the grain in which they lie with but little rotating, as a sufficiently quick and well-

lubricated needle might ever point northward, however swiftly and mazily its bearer waltzes.

To imagine the travel of these minute crystalline blocks which pass through the throes of punching undisoriented, surviving this cataclysm with ordered ranks, is an interesting mental exercise.

680. The Units Which move in Plastic Deformation are very Minute in Chemically Homogeneous Masses, but not in Chemically Heterogeneous Ones.—The absence of marked and sharp intrusions from one grain of ferrite into another indicates that, in whatever proportion the motion may be crystalline, the crystalline blocks which move as units are in general extremely minute (see end of §613). Note, for instance, the smoothness of the outlines of the ferrite grains of Fig. I of Plate 44, after great deformation by rolling. Even after the violent disturbance of punching the smoothness of these outlines persists, as shown in Figs. F and G of Plate 37. In other cases in which the ferrite is drawn out into long hairs, the individual grains retain this smoothness of outline.

In marked contrast to this absence of intrusions in masses consisting of almost pure ferrite are the abrupt intrusions which occur when the ferrite is accompanied by a harder constituent, such as cementite released by divorcing annealing, Fig. A of Plate 26, pro-eutectoid cementite, Fig. B, slag, Fig. D, or pearlite, Fig. F. In the first three of these the softer ferrite is intruded so as to fill the crevices which the deformation forms in the less ductile constituents. In a medium carbon steel Howe and Levy show what seems to be a laccolite of pearlite forced transversely across the bandings of ferrite.[1]

681. What Prevents Rotation during Slip?—What is the force which thus prevents rotation? We naturally assume that it is that same "vectorial" or directional manifestation of cohesion which originally rotated these same molecules away from the haphazard directions of the amorphous molten state into the definitely fixed directions of the crystalline state, that manifestation which we call the crystalline force for brevity.

This manifestation is even more striking than that which initially marshalls into definite orientation the molecules depositing from the liquid on the crystal. For in this initial assembling any given molecule is as free to assume the specific orientation of the crystal to which it now attaches itself as any other, so that this adoption of a definite orientation occurs without having to overcome any opposition. But it seems to be in stubborn opposition to a great force that, during plastic deformation, these crystalline blocks maintain their existing orientation, and retreat as it were in perfect order. Cohesion would, of course, make them huddle together, and perfection of huddling would lead them to retain uniformity of orientation. What is extraordinary is that in this "dance of plastic circumstance" the huddling retains this degree of perfection. This conception is made all the more difficult by the supposition that these little crystalline particles, instead of being in actual contact, are separated from each other by amorphous sheaths,

[1] *Trans. Amer. Inst. Mining Eng.*, 1914, vol. 50, Fig. 22, Plate 2, after p. 536.

however thin these may be. This thought strengthens our inference that these sheaths are extremely thin, and leads us to conjecture that their thinness is so extreme that the directive force of blocklet upon blocklet transmits itself across them.

There are, indeed, two distinct difficulties here; first, that the orientation of these blocklets remains undisturbed; second, that the mass persists in etching into rectangular pits, even though each of these blocklets is sheathed with amorphous metal. The rectangularity of the etching pits seems to me to prove this persisting of the orientation, and it is this persisting that I attempt to explain. Having no convincing explanation of why even undeformed metal which lacks amorphous sheaths etches into rectangular pits, I cannot pretend to show why the presence of these amorphous sheaths does not prevent this rectangularity of etching.

682. To sum up, twinning, both in its mechanical form which generates the Neumann lamellæ and its annealing form, is clearly crystalline, and so in a sense is intergranular motion. These three collectively contribute but slightly to plastic deformation (§622A), so that by elimination most of this is caused by that mechanism which causes slip bands. Is this mechanism crystalline or fluid? The straightness of the slip bands and more especially their crystallographic directions in the early stages of deformation argue that the deformation then is chiefly if not wholly crystalline, and the regularity of the etching pits and of the etching tint even after very great deformation argues very strongly that it has remained chiefly crystalline. On the other hand, all the many indications of important fluid motion are equivocal, because they can be accounted for either by the rounding of the profiles of the slip bands by amorphous metal, or by the integration of extremely short conjugate crystallographic lines into apparently smooth curves, or as representing the elastic stage of the deformation. The sheaths of amorphous metal encasing, on Beilby's theory, the crystalline blocklets which progressively subdivide as deformation progresses, would serve to bring about this breaking up of the paths of deformation into such undistinguishably short conjugate lines, with the resulting suggestions of fluid motion. Even the fluid motion suggested by the progressive change of etching tint in the Neumann lamellæ of Fig. A of Plate 45 appears to form only a very small fraction of the whole. The evidence that the motion is preponderatingly crystalline is so strong, and that tending to show that it is in important part fluid is so inconclusive, that we accept the former hypothesis as the more probable.

Is this crystalline motion which is the preponderating cause of slip bands, and is thus the chief mechanism of plastic deformation, twinning or slip?

The slip bands certainly resemble the Neumann bands or mechanical twins in some striking respects, in their crystallographic orientation and in their profiles. Yet there is no strong reason for regarding them as twins. On the other hand deformation, in causing slip bands, causes marked chemical and physical changes which are naturally referred to the mechanism

which has caused those slip bands. These changes are explained more readily as resulting from slip than from twinning. Moreover, the absence of traces of slip on repolishing and etching after slight deformation implies that the layers involved in the formation of those slip bands are extremely thin relatively to the deformation itself (§616). This cannot be true of twinning but might easily be of slip. Hence the inference that the slip bands form not by twinning but by slip, or in short that they deserve their name. Yet this absence of the traces of slip is in itself so surprising that suspicions lurk as to whether slip is after all the true mechanism (§526). The mechanism seems to be crystalline; twinning seems excluded; and slip is the only other crystalline mechanism thus far suggested. But why are its traces so often absent?

From out of this mass of evidence three things stand forth boldly. First, however persuasive the suggestions of important fluid motion may be, none of them is conclusive but all are compatible with exclusive crystallinity.

Second, the geometrical relations between certain sets of slip bands, the sharp intersections of certain of their treads and risers, the regularity of many etching pits even after extreme deformation, and the perfect regularity of outline and continuity which extremely thin Neumann lamellæ retain through serious twisting after their formation, are incompatible with any important degree of fluidity. Hence, on the general principle that even a great mass of inconclusive evidence must yield to a little which is conclusive, the present evidence, in my opinion, favors the belief that the deformation is preponderatingly crystalline. The fluid fraction thus far strongly indicated is not greater than could be explained by the supposed amorphous sheaths about the crystalline blocks into which deformation breaks the whole.

Third, of these kinds of crystalline motion the intergranular is slight except in the early stages of deformation, and twinning does not seem to cause any large degree of deformation. Hence by elimination we come to regard crystalline slip, as of the cards of a pack, provisionally as the chief mechanism of plastic deformation, though we cannot hold this inference firmly till we have some better explanation of the frequent failure of traces of the slip bands to appear on repolishing and etching.

CHAPTER 28

INTERGRANULAR AND TRANS-CRYSTALLINE RUPTURE

683. Introduction.—Now that we have considered in Chapters 16 and 18 the mechanism of plastic deformation in ferrite, and in Chapter 21 the plastic deformation of the conglomerates of ferrite with pearlite of which our steels consist, let us take up the logical completion of deformation by rupture. Let us consider in this chapter what preference rupture shows, in ferrite and other simple metallic substances, for each of the two possible paths, that along the grain boundaries and that across the bodies of the grains themselves. The former path leads to intergranular rupture and a granular fracture, the latter to trans-crystalline rupture, though, as we shall see in Chapter 30, the fracture while trans-crystalline in this generic sense may not be "crystalline" in the specific sense, but silky, fibrous, or porcelanic.[1] Then in the following chapter we can see what preference rupture shows for either the ferrite or the pearlite of hypo-eutectoid steel.

684. Rupture is habitually trans-crystalline in simple metallic masses, that is those which, like the ferrite of very low-carbon steel, are composed of a single substance as distinguished from conglomerates such as that of ferrite and pearlite in hypo-eutectoid steel. In calling it trans-crystalline I mean that it passes across the bodies of the grains instead of along the boundaries which separate grain from grain. We are familiar with this law through the fact that fractures are only rarely granular, and hence necessarily trans-crystalline both in iron and in other metals.

We might at first infer from this that the adhesion between adjoining grains, that is the "joint-strength," is greater than the cohesion between the particles of the grains themselves, that is the "grain-strength," somewhat as wood adheres more strongly to the glue of a joint than to itself, with the result that when it breaks rupture avoids the joint. But a little reflection shows that this inference is not necessarily true. For if the joint-strength instead of exceeding the grain-strength only just equalled it, then rupture would pass along the joints only in an insignificant fraction of cases, because in any given section the area occupied by the joints is only an insignificant fraction of the whole. But we shall see in §694 that there is an actual avoidance of the grain boundaries such as indicates that the joint-strength exceeds

[1] Early recognitions of the distinction between intergranular and trans-crystalline rupture are those of Arnold and Jefferson, *Engineering*, Feb. 7, 1896; Stead, *Journ. Iron and Steel Inst.*, 1898, No. I, vol. 53, p. 179; *idem*, 1898, No. II, vol. 54, p. 137; and *The Metallographist*, 1898, vol. 1, p. 331; *idem*, 1899, vol. 2, pp. 86–87; and Ewing and Rosenhain, *Proc. Royal Soc.*, 1899, vol. 65, p. 85.

the grain-strength. Let us return to this subject after considering the cases in which rupture does follow the joints, so that the fracture is granular and not trans-crystalline.

685. Conditions under Which the Path of Rupture is Intergranular.—The underlying condition which leads to intergranular rupture seems usually to be the presence of a weak or brittle cement between the grains themselves. A typical case, reported early by Arnold,[1] is that of an alloy of gold with bismuth which had concentrated its bismuth into its intergranular spaces as a brittle eutectic cement, rich in bismuth. When this alloy was heated to far below redness, the intergranular bismuth cement became so brittle that the whole mass could be granulated in a mortar, though the grains themselves thus separated were so malleable that they could easily be beaten out into gold leaf. So with copper alloyed with very little bismuth.

In the specimen of bismuth shown in Fig. D of Plate 41 the path of rupture, though chiefly trans-crystalline, yet is intergranular for a short distance immediately above *A*. The reason for this I do not know.

Arnold noted also a quasi intergranular fracture in certain steel castings, in which the path of rupture lay along scoriaceous envelopes surrounding the grains, as was confirmed by Stead.[2]

Stead has noted also, in addition to the pronounced trans-crystalline fracture in certain annealed sheets of very low-carbon steel, an intergranular fracture for which no clear reason appears.[3]

686. Burning Steel by Heating it above the Solidus.—The burning of steel, both hypo- and hyper-eutectoid, with its resultant granular fracture, is brought about by heating above the solidus line *AEBC* of Fig. 23, p. 130, as was shown by Stansfield,[4] who on heating a series of specimens to temperatures slightly above and below the solidus, found that those heated above this line were burnt, but that those which had not reached it were not. The way in which the matter, which starts to melt when the solidus is passed, collects in the boundaries between the austenite grains has already been shown in Figs. A and B of Plate 10. The rounding of the outlines of these grains by this incipient fusion tallies with the smoothness of the rounded surfaces of Fig. A of Plate 40.

In the presence of air, the burning of steel by heating above the solidus is likely to be aggravated by the evolution of gas along the grain boundaries, which indeed are often made thus to yawn apart.

687. Test for the Detection of Burning in Steel.—Because phosphorus lowers the melting point greatly, any phosphorus present is naturally found concentrated in this fused intergranular cement in burnt iron. On this fact

[1] Arnold and Jefferson, *Engineering*, Feb. 7, 1896; Arnold, *Fourth Report Alloys Research Comm.*, Excerpt *Proc. Inst. Mech. Eng.*, Feb. 4, 1897, discussion, p. 80; also Rosenhain, "Physical Metallurgy," 1914, p. 256.

[2] *Journ. Iron and Steel Inst.*, 1898, No. II, vol. 54, p. 140.

[3] *Idem*, 1898, No. I, vol. 53, p. 180. Here the sheets "broke up exactly like thin plates of cast iron."

[4] *Journ. Iron and Steel Inst.*, 1903, No. II, vol. 64, p. 433.

Stead bases his process for detecting burning. If the burning has gone as far as in Figs. A and B of Plate 10, so that a clear intergranular cement surrounds such molten grains, burning is evident. But even if it has not gone as far as this it may yet form molten globules rich in phosphorus, which would not occur in an unburnt specimen.[1] The enrichment of such globules in phosphorus may be detected for instance by Stead's method of covering the specimen with an acidulated alcoholic solution of chlorides of copper and magnesium.[2] Here we have a method of detecting even the less marked cases of burning.

What is true of phosphorus is true of some other impurities. They not only concentrate like it in the last solidifying parts, but on incipient fusion caused by heating above the solidus they form globules between the grains, and later a network enclosing them. Moreover, like phosphorus they may be detected by means of this same cupric reagent. Thus Stead's method serves for detecting both burning and the presence of a group of foreign elements, but not as yet for distinguishing these elements from each other.[3]

688. Other Metals Yield an Intergranular Fracture near Their Melting Points.—Rosenhain and Ewen[4] found that the purest lead, tin, gold, and bismuth yielded intergranular fractures under extremely slight stress at temperatures from 3° to 20° below their melting points.

689. Weakening the Supposed Intergranular Cement by Heating.—Rosenhain and Humfrey[5] find that, as the temperature rises and the rate of straining decreases, rupture in ferrite (low-carbon steel) tends more and more to become intergranular as in burnt iron instead of trans-crystalline as in steel under normal conditions. For instance steel of 0.106 per cent. of carbon, when strained at 932° at the rate of 0.012 in. per second, yielded more at the grain boundaries and less by slip, and hence had less prominent slip bands, than when strained in the cold; and, when the effect of this high temperature was reinforced by straining extremely slowly, it yielded almost solely along the grain boundaries and with very slight development of slip bands (their p. 260). Thus whereas when such steel is pulled in two under usual conditions in the cold there is very great drawing out of the individual polyhedral grains, with great deformation of the test piece and the abundant development of slip bands to represent that deformation, if it is pulled apart extremely slowly at 932,° it yields not by this drawing out of the grains but by their separating from each other with only little deformation either of the grains themselves or of the test piece as a whole, as shown in Fig. C of Plate

[1] *Journ. Iron and Steel Inst.*, 1915, No. I, vol. 91, p. 400, "The Detection of Burning in Steel." Stead's method is a development of the earlier ones of Heyn and of Rosenhain and Haughton (§768 and Table 37).

[2] *Idem*, "Iron, Carbon, and Phosphorus," p. 173.

[3] *Op. cit.*, p. 176, and *Engineering*, 1915, vol. 99, p. 564 (May 21). (See §781.)

[4] Rosenhain, "Physical Metallurgy," 1914, p. 260, and Rosenhain and Ewen, *Journ. Inst. Metals*, 1913, No. 2, vol. 10, pp. 127 *et seq.*

[5] *Journ. Iron and Steel Inst.*, 1913, No. I, vol. 87, p. 219, and especially pp. 256 to 266. Also Rosenhain, "Physical Metallurgy," 1914, p. 259.

42. Here fissures can be seen between the grains at some distance back from the fracture.

It is true that at 932° and all higher temperature we have to do with austenite, an allotropic form with properties radically different from those of ferrite. Yet it is very probable that, as the temperature rises still higher, the deformation becomes increasingly intergranular and decreasingly transcrystalline.

Here, and in the cases considered in the following sections, the existence of an intergranular cement is not directly observed but inferred.

690. The Path of Rupture in Steel Made Intergranular by Previous Heating in very Gently Oxidizing Gas.—Humfrey[1] finds that heating nearly or quite pure iron at any temperature above $A1$ in an extremely feebly oxidizing atmosphere, such as is created by the presence of a very little free oxygen or a little carbonic acid either in a vacuum or in nitrogen, develops great brittleness in the cold between the grains of the ferrite into which the iron transforms in cooling. These grains he distinguishes ingeniously from the somewhat smaller ones of the austenite which existed above $A3$, by the fact that it is to their boundaries that the slip bands later developed in the cold are related. The embrittlement seems to be in rough proportion to the temperature, through the range studied. Though it needs a slight oxidizing action, yet it is prevented by a stronger one. Thus if a continuous coat of oxide is formed before the heating, or is developed during heating by the presence of much carbonic acid, intergranular embrittlement does not occur. This embrittlement is always accompanied by an etching or deepening of the grain boundaries, but without tarnish of the grain bodies. At slightly lower temperatures the whole tarnishes without embrittling (his p. 89).

This brittleness is removed completely by an exposure to hydrogen at 1,000° for 1 hour, even if this is followed by a slow cooling. It is removed partly if such a heating to 1,000° lasts even 15 minutes, but not if it reaches only 700° or 820°, nor by heating to 1,000° in air for either 15 or 60 minutes followed by slow cooling.

Again, the cure is complete on vacuum heating and quenching, if this heating is for 15 minutes at 1,000°, but there is no disembrittlement if a 20-minute stay at 820° is substituted for a 1,000° exposure. If between the 1,000° vacuum heating and quenching the temperature falls slowly to 820° the cure is nearly complete, but it is slight if this intermediate cooling is carried to 700°. Finally, the cure caused by vacuum heating at 1,000° and quenching is undone by vacuum heating to 820° followed by slow cooling. The experimental evidence for some of these propositions, though admirable qualitatively, might with advantage be broadened.

691. Humfrey's explanation is that the austenite above $A1$ takes up a special low oxide which is insoluble in the ferrite, and hence is expelled by it during slow cooling through the transformation range, and like any other expelled body is thus assembled in the grain boundaries, which it embrittles.

[1] *Carnegie Memoirs, Iron and Steel Inst.*, 1912, vol. IV, p. 80.

This explanation accords with most of the facts: (1) that the brittleness is cured by an exposure to hydrogen above $A3$, fully or partly according to the time; (2) that it is caused by heating above $A3$ in a slightly imperfect vacuum or in the presence of a little carbonic acid, not enough to tarnish the surface at that temperature; (3) provided that the exposure is followed by slow cooling which would give opportunity for expulsion to the grain boundaries; but (4) not if the slow cooling is replaced by quenching.

It does not fit easily either the embrittlement at temperatures only slightly above $A1$, for here the quantity of austenite present in the nearly complete absence of carbon should be extremely small, or the preventive action of an initial continuous coat of oxide, even when all other conditions leading to embrittlement coexist. The failure of the hydrogen cure at 700° and 820° at first seems an additional obstacle, on the ground that the hydrogen ought to cause complete deoxidation. Yet the reason for this failure may be that the resultant steam escapes only very slowly and that, as long as it remains, it bars complete deoxidation of the iron.

The phenomena in general are explained as easily by the hypothesis that the solubility of the embrittling oxide increases with the temperature irrespective of the allotropic state as by Humfrey's, and this hypothesis is not open to the first of the three objections just stated, though it does not meet the last two. Hence these explanations are at best provisional.

692. Intergranular Brittleness Caused by Heating in Hydrogen.—Ruder finds that hydrogen heating, which cures the intergranular brittleness induced by Humfrey by heating in very dilute oxygen, causes like brittleness in coarse-grained silicon steel, which in its natural state has a strongly crystalline fracture. An 8-hour exposure to hydrogen at 1,400° gives it the very granular fracture shown in Fig. A of Plate 40, and makes it so brittle that a ¼-in. rod dropped 3 ft. upon a wooden floor breaks into several pieces. This effect is due to the hydrogen and not to the temperature, because it cannot be brought about by a like exposure to either air or nitrogen. On the other hand, it is invariably cured by heating in hydrogen containing hydrocarbon vapor. By alternating these treatments the cycle of embrittlement and cure can be repeated indefinitely.

There is the possibility that oxygen may be the cause of the damage here, as in Humfrey's case. His exposures at 1,400° to air must, of course, have been to a limited quantity of air, for otherwise the specimen would have been destroyed. If we suppose that the hydrogen exposures were to a stream of this gas, then this might introduce oxygen continuously as an impurity, and thus bring about gently oxidizing conditions like those which Humfrey found embrittling.

He finds that a like effect is caused by electrolytic etching with potassium dichromate. This causes an accelerating loss of weight, which in two cases amounted to 4 per cent. of the weight of the specimen at the time when the intergranular adhesion was so fully removed that the specimen began to fall apart from its own weight. As the electrolysis proceeds the grain boundaries

widen progressively, for instance to a width of 0.05 mm. at the end of 48 hours and of 0.190 mm. at the end of 168 hours.[1]

693. Cause of the Embrittlement of the Supposed Cement.—Clearly Ruder's intergranular embrittlement represents some effect of the etching reagent on something in the grain boundaries, an effect which is reversible on heating in hydrocarbon. The loss of weight and the widening of the grain boundaries suggest the removal of something, possibly carbon, first, because the embrittlement is caused by heating in hydrogen which is strongly decarburizing; second, because it is cured by adding to the hydrogen hydrocarbon vapor, which might well crack and deposit carbon; and third, because a void is evidently created at the grain boundaries, as is proved by their taking up molten copper when an embrittled specimen is immersed in it.

The 4 per cent. loss of weight in dichromate etching is evidently far more than can be accounted for by decarburization, because the specimen as a whole contains only about 0.05 per cent. of carbon. Yet boundary decarburization by the hydrogen and recarburization by the hydrocarbon vapor may play an important part.[2]

The contrasts between the conditions, both pathogenic and curative, of the embrittlement of these two different materials, the pure ferrite of Humfrey and the silicon-bearing ferrite of Ruder, suggest that the boundary metal is subject to more than one ailment, and hence may be embrittled by more than one immediate cause; or that, if there is but one ailment, it may be induced by different immediate causes under different conditions. More study is evidently needed.

694. Deformation and Rupture do not Simply Avoid Grain Boundaries but Seek Grain Centers.—Does the habitual path of rupture across the grains instead of between them represent a mere avoidance of the grain boundary as the rupture of glued wood avoids the joint because this is stronger than the wood which it joins, or is it a preference for the grain centers? If the former, then just as glued wood is as likely to break close to the glue as anywhere else, so the path of rupture would on an average lie as often close to the grain boundaries as elsewhere; if the latter, then the path of rupture would in the majority of cases lie nearer the center than the edges of the grains.

To test this I had the position of the path of rupture relatively to the grain boundaries determined in seven pieces of bismuth, in each of which a crack had been developed after the piece had been polished and etched so as to develop the grain boundaries. The procedure was as if each cracked grain was divided into five imaginary bands of equal width and running parallel to the crack, and the preference of rupture for each of these bands then

[1] W. E. Ruder, *Journ. Indust. and Eng. Chemistry*, 1913, vol. 5, p. 452. A typical analysis of the steel used is: C, 0.050; Si, 3.80 (3 to 5); Mn, 0.072; P, 0.010; and S, 0.028 per cent.

[2] The fact that the specimen as a whole showed no gain of carbon after the hydrocarbon heating is unimportant, because the boundaries form so small a fraction of the whole that their enrichment in carbon would not be disclosed by an analysis of the whole specimen.

determined. The results, condensed in Table 26, show that more than twice as much of the crack runs in the middle band as in the outer ones, and about one-half more of it in the intermediate band than in the outer ones.[1]

TABLE 26.—PREFERENCE OF RUPTURE FOR THE GRAIN CENTERS IN BISMUTH

Observer	Method	Position of the path of rupture relatively to the grain boundary, in percentage of the local grain width		
		By less than 20 per cent. of the grain width	By between 20 and 40 per cent. of the grain width	By between 40 and 50 per cent. of the grain width × 2[a]
A. G. Levy....	Length of path	22.1	31.6	46.3
H. P. Green...	Number of grains	18.25	33.65	48.10
Mean.....	.....	20.2	32.6	47.2

Note to Table 26.—[a]. The number actually observed distant from the grain boundary by between 40 and 50 per cent. has to be doubled to make it comparable with those recorded in the other columns.

This evidence, as far as it goes, tends to show that rupture seeks not only to avoid a certain limited contact region, but to keep as far as possible from the grain boundaries, as it necessarily would if the avoidance of the boundaries were due to end-support of the slipping planes. This inference, if true, would be so important that additional evidence to check this of mine is needed. It is certainly supported by Humfrey's generalization that, in ferrite, the slip bands start in the middle of the grains and work thence gradually to the (terminal?) boundaries.

The zigzagging of the crack in grain 1 of Fig. D of Plate 40 suggests in an interesting way that as the planes along which rupture passes approach the grain boundary, they receive progressively more and more end-support. Thus if we suppose that the crack propagated in the direction of the arrow, then on nearing the lower grain boundary and thus meeting increased resistance it steps to the right at right angles, and thus repeatedly reapproaches the axis of the grain. But such evidence should be received cautiously.

[1] These pieces of cast bismuth were polished, etched, and then cracked open by extending a fine sawcut made in each. They were then photographed with a magnification of 2½ diameters, and on the photographs were drawn the outlines of the grains through which the crack passed, as shown in Fig. D of Plate 40. Observer 1 then measured the total length of all these cracks which lay at a distance of less than 20 per cent. of the width of the grain from the boundary; also the total length which lay at a distance from the boundary of between 20 and 40 per cent. of width of the grain there; and finally the total length which passed at a distance from the grain boundary of between 40 and 50 per cent. of the local grain width.

Observer 2 projected the photographs and traced a greatly magnified image of them, and then counted the number of grains in which the paths of rupture lay within each of these sets of limits. The results are given in Table 26. The total number of grains crossed was 55. In order to make the results comparable they are reduced to percentages, and the numbers corresponding to the 40–50 per cent. distance are doubled.

The two observers were in ignorance of each others methods and results.

The seven specimens were from the same lot, and for all I know from the same ingot.

695. Why Deformation and Rupture Avoid the Grain Boundaries.—All the current explanations suppose greater strength of the grain boundary metal than of the rest. Those offered hitherto hold that this strength exists before the deformation, whereas that which I offer, that this strength represents the amorphization of boundary metal by the propagation of slip from the slip planes on the approach side of the boundary to the disregistering ones on the farther side, holds that this strength is developed during the course of deformation itself, as indicated by the evidence and arguments assembled in §§482 to 496.

Boundary strength, whatever its cause, should, I explain, tend to throw deformation and rupture to the grain centers, because the unit end-support for the slipping planes which that strength offers should decrease progressively from the boundaries to the centers of the grains (§487).

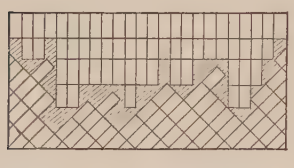

FIG. 107.—Change of orientation at grain boundaries implies spaces filled with amorphous metal. Rosenhain and Ewen.

Let us next consider the three suggested causes of boundary strength hitherto proposed: a strong amorphous form of the metal in the very joint between adjoining grains,[1] an interlocking of the metal on either side of the grain boundary,[2] and a contact region of important width stronger because of a "disturbance of the crystalline organization."[3]

696. Amorphous Boundary Filling.—If each of the little rectangles in Fig. 107 represents a crystal unit, the smallest possible mass which is compatible with crystalline structure, then at the contact of two grains differently oriented there must needs be many irregular pockets, shown hatched here, so small that no unit can fit into them.[4]

These spaces cannot be empty, because if they were the grain boundaries would be surfaces of weakness and not of strength as they actually are. Moreover, if the crystal units are of appreciable size, then fine-grained metal, because it has more boundaries, should be lighter than coarse-grained. Direct investigation of this point is needed. If the metal which fills these spaces is not crystalline it must be amorphous, and if amorphous then

[1] Rosenhain and Ewen, *Journ. Inst. Metals*, 1912, No. 2, vol. 8, p. 149. Fig. 107 shows the conditions.

[2] Rosenhain, *idem*, p. 179, and *Journ. Iron and Steel Inst.*, 1904, No. I, vol. 65, p. 358.

[3] Humfrey, *Carnegie Memoirs, Iron and Steel Inst.*, vol. V, 1913, p. 86. He holds (p. 88) that there is a "region of gradual and continuous change of orientation," and (p. 94) "a disturbance *from* the crystalline structure at the boundaries which must extend to far more than molecular dimensions." The minute discontinuous slips by means of which deformation occurs in the central parts of each grain "extend towards the (*grain*) boundaries by smaller and more numerous movements which finally become *continuous and elastic as the skin* (of the grain) *is reached*" (p. 97). (Italics mine.)

[4] This conception was, I believe, enunciated first by Osmond (*Journ. Iron and Steel Inst.*, 1911, No. II, vol. 84, pp. 61–2). It was developed by Bengough (*Journ. Inst. Metals*, 1912, No. 1, vol. 7, pp. 167–9, and No. 2, vol. 8, pp. 180–1), who inferred that the reason why rupture usually avoids the boundaries is that these are stronger than the grains because filled with Beilby's amorphous metal. The sketch Fig. 107 is due to Rosenhain and Ewen, *Journ. Inst. Metals*, 1912, No. 2, vol. 8, p. 156. These authors and Humfrey have elaborated this subject.

stronger, because it lacks the cleavages and other planes of weakness necessary to the crystalline organization.

The amorphous metal in these little pockets would not become crystalline on either heating or rest, as the supposed amorphous metal generated by the slip along the slip planes might, because each pocket is by assumption smaller than the smallest possible crystal unit. The metal here remains amorphous, not for lack of mobility enabling it to crystallize, but because each little isolated pocketful is too small to crystallize.

Even if we admit with A. G. Levy[1] that some of these spaces can be filled with crystal units which have been compressed elastically without losing their crystalline nature, others must be smaller than the size to which these units can be compressed.

Bragg's evidence tending to show that the crystal units are of less than molecular dimensions throws difficulties in the way of this whole conception of an amorphous boundary filling.

697. Is the Amorphous Boundary Filling Competent to Cause Appreciable Boundary Strength?—Interesting as the conception is, we may question its quantitative competence to explain the phenomena. The amorphous metal forms a series of extremely minute pockets, with connections tapering down to nothing, rather than a true layer, or, if you prefer, it forms a discontinuous layer. The greatest thickness of any pocket cannot exceed that of a crystal unit.[2] If, with Bragg, we hold that crystal units are of less than molecular dimensions, then this alleged amorphous metal must be so extremely thin that its mass is only an insignificant fraction of that of the grains which it parts. Its strengthening effect must be proportionally insignificant and hardly competent to give any effective end-support to the slipping planes, or to be effective in other ways to divert the path of rupture away from the grain boundaries.

If the dimensions of a crystal unit could be shown to be of a higher order of magnitude, a correspondingly greater effect of this alleged amorphous intergranular cement in offering end-support to the slip planes could be admitted. No evidence of such greater dimensions has been offered so far as I know.

If this discontinuous layer is to cause rupture to seek the grain centers by offering greater unit end-support to the lateral than to the median slip planes, it must have appreciable transverse shearing strength, because the propagation of slip across a boundary must be by transverse shear of such a layer. But what shearing strength can we reasonably attribute to such a discontinuous layer, with its thickest parts thinner than a molecule?

[1] Private communication, December 14, 1915.

[2] Compare Rosenhain and Ewen, *Journ. Inst. Metals*, 1912, No. 2, vol. 8, p. 156. "A physico-chemical unit might exist which might be many hundreds of molecules in dimension," Rosenhain, *idem*, p. 179. Rosenhain and Humfrey, *Journ. Iron and Steel Inst.*, 1913, No. I, vol. 87, p. 256. "The crystals of a pure metal are surrounded and cemented together by a *very thin* layer of the same metal in an amorphous condition" (italics mine). More lately Rosenhain appears to accept Bragg's belief as to the size of crystal units (*Trans. Faraday Soc.*, 1915, vol. 10, p. 290).

On the other hand, no matter how thin this layer it could, if weakened, lessen the boundary strength in such a way as to bring about intergranular slip, such as has been observed at high temperatures by Rosenhain and Humfrey (§689).

698. Evidence of the Existence of the Amorphous Boundary Filling.— This evidence consists in part of the widening out of the grain boundaries which etching causes.[1] Closely related to this is the greater loss by volatilization on heating *in vacuo* from fine-grained than from coarse-grained copper, silver, and zinc.[2] This excess is referred by Rosenhain and Ewen to the greater extent of the grain boundaries in the fine-grained metal, and to the more rapid volatilization of the amorphous intergranular cement than the crystalline metal. The alleged amorphous metal certainly ought to volatilize more rapidly than the crystalline, because its constituent particles lack the crystalline bond. And, indeed, Beilby[3] confirms our natural expectation that such apparently amorphous metal dissolves in solvents more rapidly than the same metal when in the crystalline state. In other words, both the vapor pressure and the solution pressure, which ought naturally to be closely related, are credibly held to be less for the amorphous state than for the crystalline.

The observation that even so ductile a metal as silver may be embrittled by long heating *in vacuo* is tendered as evidence that volatilization occurs chiefly at the grain boundaries.[4]

The intergranular fracture which lead, tin, gold, and bismuth yield at temperatures just below their melting points is used as evidence in support of the existence of this amorphous cement, though it might be due to some other cause, such as the softening of an undetected extremely thin film of a weak or brittle intergranular alloy, possibly of the metal with some gas undetected by analysis.[5]

All this evidence is extremely weak. The widening out of the grain boundaries on etching, and the greater volatilization from fine- than from coarse-grained metal, may be only one application of the general law of more rapid corrosion along the contact of two unlike substances than elsewhere, the unlikeness here consisting in a difference of orientation. The difference of potential causes a stronger attack along the contact than elsewhere, and the excess of this strength is not confined to the actual contact but continues, with progressive weakening, as we recede thence, and thus widens

[1] Rosenhain and Ewen, *op. cit.*, p. 159.

[2] *Idem*, pp. 160 *et seq.*

[3] May Lecture, *Journ. Inst. Metals*, 1911, No. 2, vol. 6, p. 6.

[4] Rosenhain and Ewen, *op. cit.*, pp. 159–161. The clear implication here is that the embrittlement is intergranular.

[5] The freedom of the outer parts of a steel ingot from gas and the appearance of blowholes at a considerable distance in from the surface suggest that, just as during solidification the carbon progressively concentrates in the last freezing parts, so dissolved gases concentrate till, at the depth at which these blowholes form, they suffice to supersaturate the metal and hence escape. By like process the gas dissolved would naturally be expelled by the grains into their boundaries like ferrite, slag, and all other foreign bodies, and might there form a weak or brittle alloy.

out the grain boundaries. Volatilization is in effect vacuum etching, and it may confidently be expected to act in general as etching by liquid solutions does.

699. The Intergranular Rupture at High Temperatures Argues for the Existence of the Amorphous Boundary Filling.—Much stronger support for the hypothesis of this amorphous boundary filling is given by Rosenhain and Humfrey in their proof that, as the temperature rises and the rate of straining decreases, rupture in ferrite and probably in other metals tends to become intergranular instead of trans-crystalline, as pointed out in §689.

The amorphous filling, because it is amorphous, is reasonably held to have in a general way the properties of undercooled and hence very viscous liquids, like barely mobile glass, hand-warmed wax, or pitch. Raising the temperature and decreasing the rate of straining certainly lessen the resistance which such liquids in general offer to distortion, and therefore ought to have this very tendency to cause the yielding to occur rather through the alleged amorphous or viscously liquid intergranular cement than across the crystalline bodies of the grains themselves. The fact that they have this effect, which on this theory they ought to have, certainly tends to support the theory.

This intergranular weakening at high temperatures is perfectly compatible with such thinness and discontinuity of the supposed amorphous intergranular layer as would deprive it of all importance as a source of boundary strength to resist the transverse shear which the propagation of slip across the boundaries implies (§697). Hence I find here nothing to imply that this supposed amorphous boundary filling is a cause of the avoidance of the grain boundaries and of the seeking of the grain centers by deformation and rupture.

Later experiments by Chappell give reason to suspect that the bodies which thus become torn apart, leaving the fissures shown in Fig. C of Plate 42, are not true grains but fragments of grains. Under this conception a grain may break up into fragments cemented together by metal which has not the orientation of the initial grain, and may perhaps be amorphous; and in the final rupture it may be these fragments which we see parting from each other.[1]

700. The Strong Contact Metal Theories. Dendritic Interlocking.—Rosenhain imagines a dendritic interlocking of the adjoining grains, such as might result from the interfering growth of the dendrites shown in Fig. 108, and holds that this should increase the boundary strength.

701. Contact Disturbance of Orientation.—Humfrey imagines a contact region bordering the boundaries themselves, and stronger because of a "disturbance of the crystalline structure caused by the difference in orientation between adjoining grains," and far wider than "molecular dimensions," in short, a region of progressively increasing amorphousness and hence great strength, as distinguished from Rosenhain's region of the interlacing of the

[1] *Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, p. 489.

pine-tree tips of two distinct systems, without loss of their individual crystalline character. Yet these two conceptions may be regarded as two modifications of the same thing, contact confusion of orientation, and they may well have closely like effects in most respects. They may therefore be tested together, by asking whether their implications correspond to the observed phenomena.

702. Evidence that There is Contact Confusion of Orientation.—This region is imagined in order to explain why slip and rupture prefer to avoid the grain boundaries. Humfrey[1] offers as evidence of its existence the curving and forking of the slip bands in ferrite as they approach those boundaries. But we have seen that this curving is explained readily as facilitating the propagation of slip along the slip planes of the grain on the other side of the boundary (§503). Moreover, this curving in ferrite fails to explain the avoid-

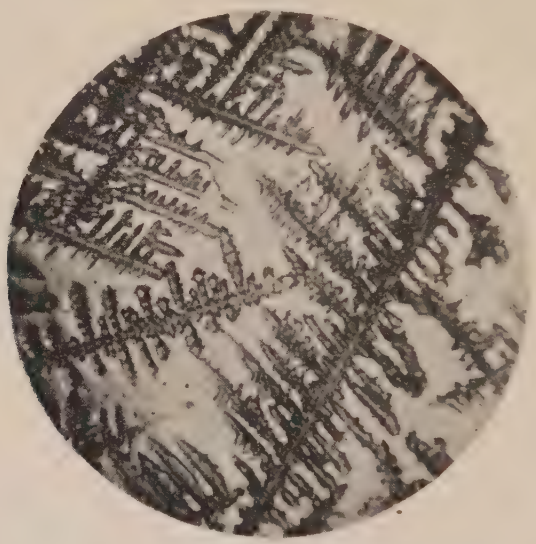

FIG. 108.—Interlacing of pine tree branches from adjoining grains in sal ammoniac, causing a boundary region of mixed orientation. Rosenhain.[2]

ance of the grain boundaries in other metals in which the slip bands are straight. Even in ferrite they are often straight, and yet the grain boundaries are avoided. Thus while the cases of curvature of slip bands do not need contact confusion to explain them, because they can readily be explained otherwise, the cases of straightness tend first to show that this contact confusion does not exist, and thus further to show that the avoidance of the grain boundaries, at least in these cases, is due to something other than confusion.

Indeed, the progressive decrease in the straightness and regularity of the slip bands as deformation progresses, not only in ferrite but in other metals, argues that any curvature of the slip bands caused by the grain boundaries is due to a resistance developed there by the deformation itself, such as the

[1] *Carnegie Memoirs, Iron and Steel Inst.*, 1913, vol. V, p. 91.
[2] *Journ. Iron and Steel Inst.*, 1904, No. 1, vol. 65, p. 360.

amorphous metal which I imagine to be generated during deformation by the disregistering propagation of slip.

703. Genesis of Contact Interlocking and Confusion.—It is true that if the growth of neighboring crystals occurred strictly as sketched in Fig. 26, then the tips of certain branches of some fir trees ought to project into the space between the branches of others in the adjoining crystals as in Fig. 108. Hence the crystal or grain boundary would be jagged or sawtoothed, and the orientation in its immediate neighborhood would be mixed, some branches following the orientation of the grain on the left and others that of the grain on the right of the boundary.

Here it is well to consider the actual type of grain boundary between the colonies of the eutectic, ledeburite. Each colony in solidifying is equivalent to a single grain, in that it is uniformly oriented throughout. In Figs. G of Plate 1 and M and N of Plate 3 we see that there is very little, if any, mixing of orientation at the contacts between neighboring colonies, which instead are separated from each other by a band of austenoid or of cementite. It is not clear that this band differs in orientation from both of the colonies between which it lies, but a band which does so differ might form in either of two ways: where the tips of branches from the two adjoining colonies come into contact, the particles on the very tips might abandon the branches on which they had been deposited and coalesce to form a band irrelated in orientation to either set of branches; or as solidification nears its end and the two systems of orientation of the two adjoining colonies strive to master the last remaining molten particles, this strife may lead these particles to refuse allegiance to either master, and instead to form a band of independent orientation. But in either way the result should be that the orientation of each colony would persist unmodified by the nearness of the other, at least up to a certain point. At this point a third orientation independent of these two competing ones may perhaps arise, and the molten still remaining and as yet unattached in now solidifying may conform to this new orientation.

But whether the sal ammoniac or the eutectic colony type is followed, there is no curving or swinging round, no progressive compromise between competing orientations. The orientation of every particle follows either the orientation of one of the adjoining grains or some third independent orientation, as in the eutectic intercolonial bands, and there is no orientation intermediate between these two or three as the case may be.

If there is a region of mixed orientation, the mixture is as that of oil and water and not of wine and water.

The absence of any confusion or progressive change in orientation here tends to increase our skepticism as to their existence elsewhere, because the conditions would seem to favor their existence as much here as elsewhere. Let us now consider three agencies which tend to efface any initial interlocking and confusion.

704. Effacement of the Interlacing and Sawtoothing of Grain Boundaries by Surface Tension.—As surface tension, opposing the crystalline force,

rounds liquid crystals, as shown in Fig. B of Plate 14, smooths the surface of liquids, and straightens the partitions between soap bubbles, so it would naturally tend to smooth and round the grain boundaries, and thus to efface any initial dendritic interlocking, confusion, or progressive change of orientation. Indeed the coalescence which has just been offered as one cause of the intercolonial band in eutectics is itself one result of surface tension, and the absence of sawtoothing between the adjoining eutectic colonies is probably in large part due to this same force.

705. Effacement of Interlacing by Grain Growth and Grain Regeneration.—The phenomenon to be explained, the avoidance of grain boundaries, often occurs in steel which has transformed a dozen times from its initial austenite structure to that of ferrite, or of ferrite plus pearlite, and back again, and between these operations has had its initial grain structure further obliterated by rolling, forging, and even wire-drawing At each of these transformations and after each of these obliterations there is the birth of a new set of grains from the ruins of the old. It is difficult to suppose that any solidificational interlocking, confusion, or progressive change in orientation at the grain contacts can survive all these destructions and regenerations.

Moreover, grain growth, if it occurs according to Rosenhain's theory,[1] would be expected to efface interlocking, progressive change in orientation, and contact confusion. This theory is that grain growth results from the

[1] *The Metallographist*, 1902, vol. 5, pp. 104–109; *Phil. Trans. Royal Soc.*, 1900, A, vol. 195, p. 279; the Author, *Proc. Amer. Soc. Testing Materials*, 1911, vol. 11, pp. 277, 379. This theory of grain growth seems incompetent to explain the extraordinarily rapid grain growth recorded by Ruder, in which "a few grains of from 3 to 5 mm." (0.12 to 0.20 in.) "formed in 3 seconds," at 1,300°, in silicon steel (*Trans. Amer. Inst. Min. Eng.*, 1913, vol. 47, p. 574). We can hardly believe that a grain-boundary migration can occur with this velocity.

Moreover, it seems incompetent to explain the perfect smoothness and sharpness of the grain boundaries developed by Humfrey in a 16-hour vacuum etching of ferrite at 1,000° (*Carnegie Memoirs, Iron and Steel Inst.*, 1912, vol. IV, p. 93, and Fig. 18, Plate XIV). If grains grow thus by the migration of their boundaries, then such migration should cover a very considerable distance in so long and high a heating. At each stage in this migration the existing boundary should be etched, and this etching at any one stage should not be effaced by the etching after the migration has passed beyond it. On the contrary, the etching of each of the successive positions of the boundary should persist, somewhat modified of course but not effaced, and hence the etched region, instead of being an extremely narrow line such as is caused by a brief etching, should be a band of a width roughly proportional to the length of the etching period. The fact that no such band but only a narrow line exists seems to me a very conclusive proof that, in this case at least, the grain growth does not occur by progressive boundary migration. It is more likely that growth occurs through the absorption of the smaller and weaker because less highly organized grains by the larger and stronger (the Author, "Life History of Network and Ferrite Grains," *Proc. Amer. Soc. Testing Materials*, 1911, vol. 11, pp. 276 *et seq.*). The rapid grain growth which occurs on heating plastically deformed metal probably represents the feeding of certain crystalline nuclei, which have escaped destruction in the deformation, on the surrounding relatively disorganized metal, and its rapidity probably reflects the great directive advantage which these nuclei have over the competing ones, an advantage which increases with the degree of deformation up to a certain point, and then in turn decreases, thus giving rise to the "critical strain" of Sauveur (§492), which may be that strain which disorganizes all but a few nuclei, thus giving those few a great directive advantage over the surrounding matter. Were the deformation less it would leave more well-organized nuclei to compete with each other; were it more it would so far destroy the organization of even the most persistent ones that they would have little advantage over the disorganized matter about them.

successive elimination of one grain boundary after another, as in the course of their migration across the metal one boundary meets or overtakes another and thus merges with it. We can hardly conceive that such a migrating grain boundary pushes in advance of itself a wide contact region either of Rosenhain's dendritic interlocking, or of confusion, or of Humfrey's progressive amorphization. On the contrary, we conceive rather that, at least as soon as this migration has proceeded to any considerable distance, any initial dendritic roughness or region of mixed orientation becomes effaced and replaced by complete regularity of structure even close to the boundary itself, so that, as any given particle deserts its initial orientation and adopts that of the new grain with which it now identifies itself, that adoption is complete and exact. This conception tallies with the brilliant smoothness of the grains, such as those shown in Fig. A of Plate 40, which we break from burnt iron or other metal which has intergranular brittleness.

But though this supposed process of grain growth would thus tend to smooth grain boundaries, and though surface tension certainly would, it by no means follows that they should succeed in making the boundaries straight. In fact they are often very strongly curved, as in Figs. A of Plate 3 and A of Plate 40.

I will now try to show that the region avoided by rupture is too wide to be explained as the direct result of any form of strengthening of the metal in and about the grain boundaries, and therefore is the more readily referred to the end-support of the slipping planes, whether through disregistry, or strength of the contact metal, or both.

706. Width of the Supposed Strong Contact Region.—We can estimate with some confidence the greatest admissable width of the contact region of alleged dendritic interlocking, confusion, and progressive departure from strict crystalhood toward complete amorphousness, of Rosenhain and Humfrey, from the width of the region in which certain natural consequences of these alleged conditions can be detected. For where these conditions exist they should cause irregularities in the slip bands, the Neumann bands, and the behavior on etching. Hence we cannot admit a width of this disturbed region greater than that in which these signs of disturbance exist.

707. The Orientation must Change at the Grain Boundaries even in Most of the Cases in Which the Direction of the Slip Bands does not.—A link essential to the reasoning is that, if change of orientation at grain boundaries necessarily causes this alleged strong contact region, it must cause it in the great majority even of those cases in which the slip bands remain straight in crossing a boundary, so that their dominant direction is not only followed closely but is the same on both sides of that boundary, as in Fig. B of Plate 24. This is because this straightness does not imply that the orientation is alike on both sides of the boundary, but only that the outcrops of the planes selected by slip are parallel in the two grains. But this parallelism of the outcrops would be accompanied in only an insignificant proportion of cases by parallelism of those planes themselves, and in the very great

majority of cases such planes, in spite of thus having parallel outcrops on the accidental plane of the microsection, would be inclined to each other in space at any angle between 0° and 180°. Only when those planes in the adjoining grains are parallel in space can the orientation be alike on both sides of the boundary.

Hence there must be a change of orientation at grain boundaries in the great majority even of the cases in which the slip bands pass straight across those boundaries.

708. Evidence of the Width of any Mixed Contact Region.—In Fig. D of Plate 22 the width of any region between *a* and *b*, and between *c* and *d*, in which the slip bands show irregularity chargeable to nearness to the boundary, cannot exceed some 2 per cent. of their length, or 0.000,04 in. Here as in some other cases, there is indeed a narrow light selvage between the grains, but occupying only an insignificant fraction of the width of the grain.

Note how the slip bands in grain I of Fig. E of Plate 23 show no sign of disturbance right up to the boundary of grain II, and how some of those in grains I, II, and III of Fig. D of the same plate remain unchanged right up to the right-hand terminal boundary. If there is any irregularity of the slip bands due to proximity to the boundaries it must in many of these cases be less than 0.01 in. on this magnification, or less than 0.000,02 in. in true size. So with the slip bands of grains II, III, and IV of Fig. C of Plate 23, and with many of the grains in Figs. F of Plate 16, F of Plate 22, B of Plate 31, and A and E of Plate 39. The evidence given by the figures of Plate 23 is the more direct because here the perturbing effect of the amorphous polished surface layer has been removed by the etching of these surfaces before the slip bands were created.

We have already seen in §706 reason to hold that any region in which slip bands behave abnormally, if it exists at all, is extremely narrow. Moreover, we have seen in §604 that the Neumann bands, both in ferrite and in other metals, show no sign of irregularity in approaching grain boundaries. If there is any region of contact irregularity in those of Fig. D of Plate 41 it is not as wide as 0.000,4 in.

709. The evidence given by the etching figures tallies completely with that given by the slip and Neumann bands. Note that the regularity of the etching pits in Fig. D of Plate 18, and that of the tile-like etching figures in Fig. E of Plate 40, is the same alongside the grain boundaries as farther off. The slightness of the influence of the grain boundaries on the regularity of the etching pits is illustrated by the absence of any visible change in that regularity where the etching pits actually cross the boundaries, spanning them and reaching from one grain into another, as in Figs. H of Plate 17 and D of Plate 18. Even on magnification of 1,000 diameters no change in that regularity at the boundaries is visible in a micrograph of Ewing and Rosenhain, in which a large etching pit lies in three grains.[1]

[1] *Phil. Trans. Royal Soc.*, 1900, A, vol. 193, p. 357 and Fig. 5, Plate 16.

710. The width of the avoided region is of a higher order of magnitude than that of the supposed strong contact region, strengthened by interlocking or disturbance of orientation. Irregularity of slip bands or Neumann bands in approaching a grain boundary might be due to some cause other than irregularity of orientation, but their straightness seems to imply that the orientation is regular. The evidence in §708 may thus be interpreted as meaning that, if there is any region of irregularity, it does not exceed 0.000,02 in. in one case, 0.000,04 in a second, and 0.000,4 in a third, that of bismuth, in which the width of the region which rupture prefers to avoid is at least 0.1 in. Thus in this case the width of the avoided region is at least 250 times, if not indeed 5,000 times, as great as that in which the slip and Neumann bands permit us to assume that irregularity of orientation exists.

How wide the avoided region is in ferrite is not known, but we naturally infer that it is of a higher order of magnitude than the extremely narrow region in which any signs of confusion of orientation exist.

711. The Change of Deformation at High Temperatures from the Transcrystalline to the Intergranular Type Indecisive as between the Hypothesis of an Initial Boundary Filling and that of Progressive Boundary Amorphization.—I held in §697 that the supposed intergranular amorphous cement was competent to explain this change, though it was incompetent to explain the avoidance of the grain boundaries by deformation and rupture in the cold. The reasons which I gave for that incompetence were, first that so thin and discontinuous a layer could have no important transverse shearing strength, and second that the evidence tends to show that the avoidance of the grain boundaries is due to something generated in the deformation itself, and not to something which existed before the deformation.

Hence if the observed avoidance of grain boundaries is due to my accumulation of amorphous metal there, that accumulation must be of a higher order of thickness than is admissible for the supposed pre-existing amorphous layer. But on high heating, though the intergranular amorphous layer might well persist because its amorphousness is due to the fact that spaces smaller than a crystal unit, and filled with iron, exist in the grain boundaries, and that in these spaces the iron must be amorphous because their smallness prevents it from crystallizing, yet my supposed amorphous accumulation could not persist because its persistence in the cold is due solely to the lack of mobility, permitting recrystallization. Instead it must recrystallize as soon as the temperature becomes high enough to give that mobility, which it would certainly do before it passed far above the transformation range.

This recrystallization of my amorphous contact metal would remove the obstacle which, on my hypothesis, diverts deformation in the cold from the boundaries to the grain centers, and would thus permit the deformation to resume its initial intergranular path. This change of deformation from transcrystalline to intergranular at high temperatures is thus a natural consequence of each of these reciprocally consistent hypotheses. Yet there is the difference that it gives a direct support to the hypothesis of an intergranu-

lar amorphous cement which it does not give my hypothesis, in that at these high temperatures the tendency to intergranular motion is apparently actually much greater that it is even at the very incipiency of deformation in the cold. Thus the high temperature does more than remove the obstacle to intergranular deformation which my amorphous metal about the boundaries is supposed to cause.

712. Summary of the Discussion of the Causes of Grain Boundary Avoidance by Deformation and Rupture.—The causes which we have considered consist of four imagined forms of boundary strength:

(A) Three supposed to exist before deformation, viz.:

(1) Strong amorphous boundary filling (Osmond, Bengough, Rosenhain *et al.* (§696).

(2) Dendritic interlacing of crystals across the boundaries (Rosenhain, §700).

(3) Progressive confusion of crystalline structure as the boundaries are approached (Humfrey, §701).

(B) One supposed to be developed during deformation, viz.:

(4) Progressive accumulation of amorphous metal about the boundaries through the break-up of crystalline metal caused by the disregistry of the slip planes in adjoining grains (the Author, §481).

Antecedently it is extremely probable that the first cause exists, and it seems to me probable that the fourth does, but not probable that either the second or third does habitually at the time of deformation, because even if we concede their existence at the moment of solidification, they would probable be effaced by the processes of transformation, grain growth, and, except in the case of castings, of mechanical kneading.

The competence of the first to explain the avoidance is very doubtful, because shearing strength, the property to which we attribute the avoidance, could hardly be given in an appreciable degree by an intergranular layer at once discontinuous and probably of only about molecular thickness (§697). If the existence of the three other causes were admitted, their competence to explain the avoidance might well be conceded.

The causes of class (A) are opposed, and cause (4) is favored by the evidence tending, however incompletely, to show that both the avoidance itself and the boundary strength increase as the deformation progresses.

That the avoidance itself thus increases is indicated by the contrast between the uplift of whole grains along their contact with their neighbors early in the deformation (§488) and the complete concentration of rupture, which is the end of deformation, in the grain centers, as far as possible from the grain boundaries (§694).

That the boundary strength thus increases is indicated by this same contrast; by the failure of a Turner scratch to narrow on crossing grain boundaries (§491); by the greater effect of grain coarsening by overheating on the tensile strength than on the Brinell hardness (§494); and by the progressive increase in the excess of resistance to deformation in fine- over that

in coarse-grained steel, not only from the yield point, through the Turner hardness, to the maximum or tensile strength (§490) but also with the degree of deformation which precedes the Brinell test (§495).

The curvature and irregularity of the slip bands as they approach the terminal boundaries have been offered as evidence of the third cause of boundary strength, and they apply equally to the second. They carry little weight, first, because if they could be caused by either of these two forms of boundary strength, then they could be caused by cause (4) as well, cause (1) being too weak; second, because they can be explained as the means of facilitating slip across the boundary to a disregistering set of planes on the other side; third, because if they were due to any form of boundary strength, it should be one which, like the fourth cause, increases with the deformation, because this irregularity, while in the first stages of deformation it may be too slight to be detected, yet increases greatly as deformation proceeds; and fourth, because the regularity of the slip and Neumann bands in some metals prior to deformation extends to the grain boundaries through the region in which the slip bands curve after deformation, and because the regularity of the etching pits and figures, even of the pits which span the boundaries, both before and after deformation, extends even on magnification of 1,000 diameters quite to the boundary. The perfect smoothness and the brilliancy of intergranular fractures tallies with this regularity of the deformation and etching phenomena even at the boundaries, and thus further opposes the second and third causes.

The deeper etching at grain boundaries than elsewhere, and the greater loss by volatilization of fine- than of coarse-grained metal are not good evidence of the existence of the first cause, because they would naturally result from the difference of potential at those boundaries, which are more extensive in fine than in coarse metal.

The cessation of boundary avoidance at high temperatures would naturally result from either the first or the fourth cause, because the softening of the amorphous boundary filling would divert rupture to this intergranular path, and because the return of the strong amorphous metal about the boundaries to the weak crystalline state would remove the cause which in the cold diverts deformation from the boundaries toward the grain centers. But it is not explained by either the second or the third cause.

To sum up, the first cause may well exist, but it seems too weak to explain the diversion of deformation from the grain boundaries.

The second and third probably do not exist at the time of the diversion to be explained, and are opposed by the smoothness of granular fractures, by the regularity of the slip bands and Neumann bands caused by incipient deformation, and by that of the etching figures and pits both before and after deformation. Moreover, they are incompetent to explain why rupture becomes intergranular at high temperatures.

All the causes of class (A) are opposed by the evidence tending to show that the boundary strength increases with the progress of deformation.

To the fourth cause I find no objection thus far, for it seems antecedently probable, compatible with all the evidence, and indeed almost needed to explain much of it. Of course, like every other newly enunciated hypothesis it must be received with caution till it has been tested by experiment and discussion. If it falls, may its ruin serve the builder of a better.

CHAPTER 29

RELATIVE PREFERENCE OF THE PATH OF RUPTURE FOR FERRITE AND PEARLITE

713. Introduction.—Let us ask through which of the microscopic constituents, pearlite, pro-eutectoid ferrite, and pro-eutectoid cementite, the path of rupture prefers to pass under various conditions, taking up in Chapter 30 the appearance of the fracture, that is, of the path of rupture laid bare and seen in the broken piece, and in Chapters 31 and 32 the subject of fiber.

714. Important Principles.—The explanations of the path of rupture given in this chapter are based in large part on two important conceptions. The first is that the trans-crystalline strength of the mass increases with the fineness of the grain size, because with this fineness the number of grain boundaries to be crossed by slip and rupture increases, and because the resistance offered by these boundaries to deformation and rupture is greater than that of the metal within the grains themselves, as explained in §§479 to 497. The second is that, though a momentary and local loss of strength must needs result from each fraction of slip along any slip plane, yet this is quickly outweighed by the strengthening which follows, as is familiar through the strengthening effects of all forms of plastic deformation. In considering these latter facts it is convenient to use Beilby's theory that, at the instant of slip, the contact or slipping metal along the slipping surfaces passes into a weak mobile state, from which it passes rapidly into an amorphous state stronger than the initial and crystalline state, as explained in §§508 *et seq.* All explanations which, like these, have not stood the test of discussion, must be received cautiously. They are offered as first attempts to clarify the subject.

We should recognize at the outset that great variations in the paths of deformation and rupture are to be expected, because these are the paths of least resistance, and because the resistance varies not only with the time which elapses after slip, the mobile metal generated along the slip planes setting and strengthening progressively, but also with the direction, being least along the slip and parting planes. The path actually followed must depend on the relation of the externally imposed stresses to the relative resistance of the metal in different directions. The metal is composed of polyhedral grains each of which is a perfect crystal internally, with distinct planes of weakness. We have, indeed, at least two and usually three if not four internal surfaces of weakness, the cleavages proper which every grain must needs have; the slip planes which may not be true cleavage planes; the partings between ferrite and pearlite, ferrite and slag, etc.; and in some cases the grain boundaries.

Even if the geometrical distribution of the stress were accurate, whether it caused movement along one or along another of these surfaces of weakness would depend on the quantitative relation between its components along the different ones. Here a slip plane is nearly parallel to the direction of maximum stress, and hence is the path selected by rupture; there it is not a slip plane but a parting which lies nearest to this direction of maximum stress, and accordingly rupture selects the parting; still elsewhere a grain boundary offers the easiest accommodation to rupture, happening to lie nearly parallel to the direction of maximum stress while all the slip planes and partings happen to make large angles with this direction.

Were the stress distributed initially with perfect regularity, that regularity would disappear with the first plastic deformation, for this deformation itself is distributed very irregularly both because the various planes in each grain vary so greatly in their resistance, and because of the random relations between the orientations of adjoining grains. Moreover, these irregularities themselves are forever shifting kaleidoscopically with every slip along every yielding surface.

Again, as between ferrite and pearlite, even if such attraction as ferrite has for the path of rupture were not opposed by the counter attraction of the pearlite because of its brittleness, that path could not follow the ferrite exclusively, for instance in a steel of 0.80 per cent. of carbon, for the simple reason that here there is only 11 per cent. of ferrite mixed with 89 per cent. of pearlite, so that rupture cannot cross the mass without traversing much pearlite. If rupture has a less total resistance to overcome by a short cut across a strong constituent than by a roundabout path along a weaker one, it will cut through the stronger.

In a steel of 0.10 per cent. of carbon the conditions are reversed. If it were true that rupture tended to pass through pearlite by preference, that tendency could not fully govern the path of rupture across this steel, and confine it to the 11 per cent. of pearlite present.

Hence all that we can look for is the path preferred by rupture, and we may find first, second, and third preferences, and variations of preference with the varying conditions of rupture.

These conceptions should be clear if this chapter is to be understood.

715. The Path of Rupture in Hypo-eutectoid Steel.—Though we might at first expect the path of rupture to show a strong preference for avoiding pearlite and passing through ferrite, both because the ferrite is less than half as strong as the pearlite[1] and because the earlier stages of plastic deformation occur chiefly in the ferrite, (§531) yet we quickly see a reason why rupture should certainly start in the pearlite, and thereafter should be attracted strongly to it in propagating across the mass.

[1] The strength of ferrite, as we know it in the almost pure A.R.M. steel, is about 40,000 to 50,000 lb. per square inch. This and other data lead me to infer that the strength of pure ferrite when not strengthened by plastic deformation, is about 42,000 lb. per square inch. The strength of pearlite as we know it in eutectoid steel of 0.90 per cent. of carbon varies very greatly with the thermal and mechanical treatment. Its usual limits are between 100,000 and 130,000 lb. per square inch.

In hypo-eutectoid steel the pearlite forms a continuous skeleton and the ferrite another, "nerve with nerve entwined." Under tensile stress, for instance, they are in the position of a wire rope of which certain of the steel strands are replaced by strands of very flexible india rubber. When such a rope is strained the steel will clearly break before the rubber, simply because, though both must elongate exactly alike, the power of the steel to elongate without rupture is far less than that of the rubber. Here rupture will start not in the weaker but in the less extensible member; and so for like reason it should in the pearlite of hypo-eutectoid steel.[1] If it were possible to separate the ferrite skeleton from that of the pearlite, and then, setting them side by side, to strain them tensilely and equally, the pearlite would clearly break long before the existing extension had exhausted the extensibility of the ferrite. If this is true when they are thus isolated in the imagination, it ought to be equally true when they are intertwined as in our hypo-eutectoid steel, this intertwining here compelling them to elongate exactly alike.

This principle is illustrated by the familiar "cupping rupture" of segregated wire rods, in wire-drawing. The segregated axial parts are certainly stronger than the outer parts, because they are richer in carbon. Nevertheless, because they are less ductile, they break under a degree of deformation which the outer and more ductile layers undergo unbroken. So with the "split head" rupture of segregated rails. The transverse flow of the metal at the upper surface of the rail, under the spreading action of the wheel-load, may imply greater deformation than the segregated and certainly stronger metal below can endure without breaking. Thus a crack starts in the stronger segregated metal, whence it propagates downward and longitudinally.

As rupture propagates from an initial crack started in the pearlite, across the intimate mixture of ferrite with pearlite of which hypo-eutectoid steel consists, it is exposed simultaneously to two opposite attractions. The inextensibility of the pearlite offers one attraction, and the weakness of the ferrite a counter one. The path actually chosen is of course influenced by both these attractions; but the ratio between these influences must vary with the conditions.

This inference that rupture starts in the pearlite by preference, at least when the conditions lead to great deformation, is supported by the occurrence of fissures in the pearlite near the path of rupture but not directly connected with it, as observed by Rosenhain.[2]

Because it is the slightness of the extensibility of the pearlite that attracts the path of rupture, that attraction ought to be especially strong when the conditions are such as to draw heavily on that extensibility, that is when the

[1] That this integral union of a stronger and less ductile with a weaker and more ductile part of a metallic mass will thus lead to the breaking of the former and the spreading of the crack thus started across the latter I showed in the first edition of this work, 1890, p. 212, col. 2. Rosenhain applies corresponding considerations to this present case of the mixture of ferrite and pearlite in hypo-eutectoid steel (*Journ. Iron and Steel Inst.*, 1906, No. II, vol. 70, pp. 220–221).

[2] *Op. cit.*, p. 221.

deformation is great; and because it is the weakness of the ferrite that attracts that path, that attraction will decrease as the weakness of the ferrite is decreased, and this too should occur when the deformation is great, as §723 will explain.

716. The Preference of the Path of Rupture for Ferrite is Reported to be Inversely as the Degree of Deformation.—This enunciation agrees with the inferences just drawn. The reports of several investigators when compared point to such a law, the reported preference for ferrite being especially marked in fatigue testing, in the impact testing of notched bars, and indeed under most conditions of shock, and often when rupture is caused by the propagation of a crack.

TABLE 27.—THE PATH OF RUPTURE IN HYPO-EUTECTOID STEEL

It is chiefly through ferrite if rupture is accompanied by little deformation, but through ferrite and pearlite indiscriminately if there is great deformation.

No.	Observer	Condition leading to rupture	Path of rupture
		Great deformation	
1	Seaton and Jude	Tensile test	Indiscriminate.
2	Rosenhain	Tensile test	Indiscriminate.
		Intermediate deformation	
3	Seaton and Jude	Repeated impact on notched bars	Through ferrite chiefly.
4	Rosenhain	Bending notched bars	Transverse, indiscriminate. Longitudinal, ferrite chiefly.
5	"	Bending back and forth	Chiefly ferrite, but at edges runs indiscriminaetly.
6	Robin	Impact on end of a cylinder	Begins in pearlite, usually propagates across pearlite and ferrite.
		Slight deformation	
7	Howard	Quiescent bending	Between columnar grains in ingot.
8	"	Static compression	Ferrite network (rail steel).
9	Howard and Fay	Bent statically	Ferrite network and MnS.
10	Seaton and Jude	Fatigue, rotating under flexure	Ferrite.
11	Stanton and Bairstow	Fatigue, reversals of direct stress	Ferrite network.
12	" " "	Fatigue, reversals of impacts	Ferrite network.
13	Rosenhain	Single impact on notched bar	Ferrite chiefly.
14	Ziegler	Shock	Ferrite network.
15	Rosenhain	Crack	Ferrite by preference.
16	"	"	Ferrite and pearlite indiscriminately.
17	Belaiew	.....	Ferrite network.

The page numbers given in these notes are those of the original memoirs to which reference is here made, and not those of this volume.

Note to Table 27.—1. Seaton and Jude, tensile, *Proc. Inst. Mech. Eng.;* through *Engineer-*

ing, 1904, vol. 78, p. 767. This is based on my own inspection of the micrographs of Seaton and Jude, who do not seem to have noticed that the path of rupture under quiescent tensile test passes indiscriminately through ferrite and pearlite, forming an exception to their general inference that it passes chiefly through ferrite.

2. Rosenhain, steel of 0.15 per cent. of carbon, *Journ. Iron and Steel Inst.*, 1906, No. II, vol. 70, p. 189. Rupture passes indiscriminately through ferrite and pearlite, whether lamellar, granular, or sorbitic (p. 221).

3. Steel of 0.30 per cent. of carbon, and hence with 67 per cent. of pro-eutectoid ferrite and 33 per cent. of pearlite. Sixty-four per cent. of the length of the path of rupture was through ferrite, 17 per cent. through pearlite, and 19 per cent. at the junction of ferrite with pearlite (p. 766).

4. Steel of 0.15 per cent. carbon. In those parts of the fracture which run transversely of the test piece, so that the conditions resemble those of tensile testing, rupture passes indiscriminately through ferrite and pearlite. In those parts which run lengthwise of the test piece, rupture passes chiefly through the ferrite, the tendency to pass here being favored by the longitudinal slag threads, lying as usual in the ferrite (p. 225).

5. Three-quarter-inch square bars of steel of 0.15 per cent. carbon were broken by bending back and forth through an angle of 180° twice (p. 225).

6. Robin, "Etude Micrographique sur la fragilité des aciers a l'ecrasement," *Rev. de Metallurgie, Memoires*, 1911, vol. 8, p. 448. The deformation was given by means of blows on one end of a cylinder firmly supported on its lower end. I do not infer with any confidence from his language how great the degree of deformation was, nor whether rupture showed a preference for either the ferrite or the pearlite.

7. Howard, *Proc. Amer. Soc. Testing Materials*, 1909, vol. 9, p. 324. Pieces cut from rail-steel ingots.

8. Howard, *loc. cit.* Pieces cut from a coarse-grained bloom after an early pass through the rail mill, tested in compression. Also *Iron Age*, 1909, vol. 84, p. 91.

9. Howard, "Some Results of the Tests of Steel Rails in Progress at Watertown Arsenal," *Proc. Amer. Soc. Testing Materials*, 1908, vol. 8, p. 53, and Fay, "A Microscopic Investigation of Broken Steel Rails," *idem*, p. 74. Many cracks were found along masses of manganese sulphide or slag, and along the ferrite network of rail steel, especially when this network contained sulphide or slag. The cracks which formed spontaneously in service in the martensitic spots in rail heads, extending below into the unhardened metal, were found lined with manganese sulphide. This would naturally both facilitate cracking and tend to fix the exact position of any cracks which form.

10. *Engineering*, 1904, vol. 78, p. 731.

11. Stanton and Bairstow, "On the Resistance of Iron and Steel to Reversals of Direct Stress," *Coll. Researches Nat. Phys. Laboratory*, 1907, vol. 3, p. 36 and Plate 5 (from *Proc. Inst. Civil Engineers*). Direct tensile and compressive stresses in the ratio of 7 of the former to 5 of the latter were reversed 800 times per minute. When a well-marked network structure had been created by air-cooling bars of steel of 0.64 and 0.44 per cent. of carbon from 1,000°, rupture passed chiefly along the relatively narrow ferrite network, through avoiding the pearlite which should have occupied some 71 per cent. of the total in the 0.64 per cent. carbon steel.

12. "The Resistance of Materials to Impact," Stanton and Bairstow, *Engineering*, Nov. 27, 1908, p. 734, from *Proc. Inst. Mech. Engineers*, Nov. 20, 1908. In a steel of 0.60 per cent. of carbon, in which a coarse network structure had been set up by slow cooling from 1,000°, the crack caused by a repetition of impacts alternately in tension and in compression passes in some fields "unmistakably through the ferrite" network, but in others "chiefly through the pearlite"—"although the greatest number of observations seem to point to the ferrite as the weakest constituent." In one of the two photographs which they publish the great excess of pearlite is clearly avoided and rupture passes along the narrow ferrite network: in the other it passes in part through each constituent, though the ratio of its path through ferrite to that through pearlite is much greater than the ratio of ferrite to pearlite in the specimen. This is what should be expected in view of the competition of two tendencies, one to carry rupture through the weaker ferrite, the other to carry it by the shortest path.

13. Steel of 0.15 per cent. carbon (p. 224).

14. Ziegler, "Sur la crystallisation du fer alpha," *Rev. de Metallurgie, Memoires*, 1911, vol. 8, p. 664. The steel was very coarse grained, with apparently about 0.60 per cent. of carbon, though the content cannot be estimated closely. The conditions of shock are not indicated. The

slag content of many of his specimens was very abnormally great. This of course would tend to confine rupture to the ferrite network for this contains the slag.

15. Rosenhain, "The Explosion of A Thermal Storage Drum at Greenwich," *Engineering*, 1908, vol. 85, p. 114. The steel contained 0.181 per cent. of carbon. It failed through the propagation of a crack referred to excessive cold deformation. Rupture passed "as far as possible through the ferrite."

16. Rosenhain, *Engineering*, 1908, vol. 86, p. 341. A gun tube containing 0.41 per cent. of carbon, and of good microstructure, developed internal cracks and "fissured" completely after abnormally short life, apparently because of excessive slag enclosures. The path of rupture could not be determined with confidence, but seemed to cross ferrite and pearlite indiscriminately

TABLE 27A.—PATH OF RUPTURE IN VARIOUS IRON PRODUCTS

No.	Observer	Material	Carbon content	Path of rupture
1	Common observation	Burnt iron and steel	.....	Intergranular.
2	Robin	1 per cent. phosphorus steel	.....	Intergranular.
3	Stead	Divorced low-carbon steel	0.04	Through ferrite.
4	Stead, G. W. Sargent, Author	Hyper-eutectoid steel with cementite network	1.38 and 1.00	Along the cementite network.
5	Howe and Levy	Manganese steel with cementite network	.....	Along the cementite network.
6	Osmond	Martensite	.....	Hardening cracks prefer the boundaries of the martensite (austenite?) grains.
7	Common observation	Gray cast iron	.....	Along the graphite plates.

Note to Table 27A.—Robin, "Etude Micrographique sur la fragilité des aciers a l'ecrasement," *Rev. Metallurgie, Mem.*, 1911, vol. 8, p. 445. "Undephosphorized steel," probably the blown but not afterblown basic Bessemer metal, containing 1 per cent. of phosphorus, has very marked intergranular brittleness. At temperatures above 650° the grains forge well, but at 500° "forging determines a rapid separation of all the grains."

3. Stead, *Fifth Rept. Alloys Research Committee*, 1899, p. 75. This was a soft Swedish steel of 0.04 per cent. of carbon, the pearlite of which had been completely divorced, leaving its cementite in the form of atolls; "the fracture passed right through the middle of the grains, following the cleavage planes." Rupture was caused by a blow on the specimen which was resting on a "hollow dished-out block of metal."

4. Stead, *idem.* Steel of 1.38 per cent. of carbon was cooled from about 1,300° very slowly in a ball of molten slag. Here rupture, caused in the same way as in 3, passed along the white cementite network like that of Fig. E, Plate 41.

Also G. W. Sargent, *Trans. Amer. Inst. Min. Eng.*, 1901, vol. 31, p. 304. Steel of 1 per cent. of carbon, furnace cooled from various temperatures from 650° to 1,150°. "Where crystallization had occurred," probably meaning when a cementite network structure formed, "the breaks followed the crystal faces." This case corresponds to "moderate deformation" in spite of its representing quiescent tensile testing, because in the presence of the cementite network the deformation would naturally be slight, the greatest reduction of area on rupture under these conditions being 8.5 per cent., on furnace cooling from 915°.

See also Fig. E, Plate 41. Rupture caused by tensile testing.

5. "Are the Deformation Lines in Manganese Steel Twins or Slip Bands?" *Trans. Amer. Inst. Min. Eng.*, 1915 (Bulletin No. 99, March, 1915). Rupture caused by bending manganese steel in which an abundant network of pro-eutectoid cementite had been generated by heating to 575° after quenching.

6. Osmond, "Methode generale pour l'analyse micrographique des aciers au carbone," *Bulletin Soc. d'Encouragement*, May, 1895, and Osmond, Stead, and Sidney, "Micrographic Analysis of Metals," Griffin, London, 1913, p. 221. Steels of 0.45 and of 1.24 per cent. of carbon, quenched from various temperatures. The hardening cracks tend to follow the outlines of the structural grains of martensite, but may be diverted from them by quenching stresses, and may follow along the martensite needles.

This principle is illustrated by the cases collected in Table 27, which are divided up according to the degree of deformation. Note that in the tensile tests, both of Seaton and Jude and of Rosenhain, in which the deformation is great, rupture passed indiscriminately through ferrite and pearlite, whereas in the fatigue tests of the former investigators, and of Stanton and Bairstow, in which the deformation is extremely slight, it showed a preference for the ferrite.

The clearest illustration is given by Rosenhain's cases Nos. 4 and 5. In No. 4 the fracture runs in part transversely, in part longitudinally. Where it runs transversely the conditions resemble those of tensile testing probably with great local deformation, and here rupture passes indiscriminately through ferrite and pearlite. Where it runs longitudinally its course is straight, and with little local deformation; and accordingly here it passes chiefly through the ferrite, being aided no doubt by the presence of longitudinal slag masses in the ferrite, and by the longitudinal banding of the ferrite and pearlite themselves, shown in Plate 43, as in so many other cases. Again, when rupture is caused by bending back and forth, as in No. 5 of Table 27, it passes indiscriminately through ferrite and pearlite in the outer parts of the test piece where the deformation is great, but it passes chiefly through the ferrite in the middle parts of the piece, where the deformation is relatively slight, because it is brought about by means of a crack propagated inward from the outer layers.

717. The Effectiveness of the Preference for Ferrite is Inversely as the Carbon Content. My Experiments.—That this must be true is almost self-evident. In steel of only 0.15 per cent. of carbon, such as Rosenhain used in his experiments of Table 27, there is 83 per cent. of ferrite and only 17 per cent. of pearlite, and here rupture can follow its preference for ferrite relatively easily, without having to lengthen its path materially in order to avoid pearlite. On the other hand, a steel of 0.59 per cent. of carbon consists of about two-thirds pearlite and one-third ferrite, and the avoidance of this large quantity of pearlite must usually imply a long detour. Indeed the difficulty of avoiding pearlite is usually much greater than these numbers imply, because in such a steel much of the ferrite is usually closely intermixed with the pearlite, and only a little is concentrated in masses large enough to attract rupture by their smaller unit strength. Note for instance in Fig. C of Plate 41, which shows a crack lengthened in steel of 0.59 per cent. of carbon by gentle tapping, how small a part of the field is occupied by compact masses of the white ferrite. Rupture, in taking the path of least resistance, is governed by the relative lengths of competing paths as well as by their unit strength.

The greater ease with which pearlite is avoided in lower than in higher carbon steel is seen on comparing Fig. F of Plate 42 of a steel of 0.214 per cent. of carbon with Fig. C of Plate 41[1] of one of 0.59 per cent. In the former the

[1] The specimens shown in Figs. C and E of Plate 41 and A, B, and F of Plate 42 I had broken part way across by starting a crack by means of a knife-edge file and spreading it by slight blows.

whole of the left-hand side of the crack passes through ferrite. The broken lines from *a* to *b* and from *c* to *d* indicate how rupture, by a slight detour, has avoided passing through the dark pearlite islands below them. So the avoidance of the pearlite masses between *e* and *f* has implied only a very slight lengthening of the path of rupture.

In Fig. C of Plate 41, on the other hand, the avoidance of pearlite would have been much more difficult. Here the general course of rupture is from *A* toward *C*. On reaching *A*, rupture gratifies its preference for ferrite by stepping sharply to the left as far as *B* through the ferrite network. But it soon resumes its main direction, and shoots by zigzag steps to *C* across the kernel or mesh, mainly of pearlite though with some ferrite mixed through it in masses so small that they cannot be recognized here. To have travelled from *A* to *C* wholly along the ferrite network, by passing either to the right from *A* or further to the left from *B*, would have lengthened the path so greatly that its total resistance, in spite of the weakness of the ferrite network, would have been greater than that of the shorter path actually followed across the pearlite kernel.

718. The Effectiveness of the Preference for Ferrite is Probably Inversely as the Network Size.—In the network steel which we have just considered in Fig. C of Plate 41 the difficulty of following the preference for ferrite is probably increased by the coarseness of the ferrite network itself, which not only increases the length of the detour which a path solely through ferrite would imply, but probably also lessens the total quantity of ferrite which the network contains, the remainder lying within the pearlite kernels. This network, of course, represents the outline of the austenite grains out of which the existing ferrite plus pearlite structure has formed in cooling through the transformation range, and it is, therefore, the coarser, the higher and the longer that austenite has been heated.

As the austenite grains increase in size, they have correspondingly greater difficulty in expelling to their peripheries the pro-eutectoid ferrite which they generate, and thus in distributing it as a network about themselves. With this difficulty increases the proportion of that ferrite which remains unexpelled.[1] When the austenite grains are very large, as in greatly overheated steel, the unexpelled ferrite becomes increasingly prone to assemble in the Widmanstättian or lattice figuring shown in A and B of Plate 28, and B of Plate 42. It is in accordance with these conceptions that this Widmanstättian structure is very pronounced in meterorites, in which the sojourn at a very high temperature has been so long that the whole of the austenite

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The surface was then polished, and the crack was next spread a little further by means of light taps of a hammer. The surface was then repolished very lightly and etched. In this way the ruffling of the surface by the deformation was made very small, and at the same time the last polishing could be made so slight that it did not round off the edges of the crack seriously. In this way one can study the path of rupture without the need of copper-plating it.

[1] This generalization I owe to Stead, *Proc. Cleveland Inst. Engineers*, Session 1913–14, p. 250.

commonly forms a single grain, with correspondingly great distances for the ferrite to travel in order to reach the outside.

Clearly the presence of all this internal weak ferrite within the kernels, which thus are a mixture of ferrite plus pearlite, in itself facilitates the passage of rupture across those kernels, and detracts correspondingly from the attractiveness of the longer path along the ferrite network.

Here arises a difficulty in demonstrating by this cracking method the preference of rupture for ferrite. It is only in a relatively coarse-grained material that the relation of the path of rupture to the ferrite and pearlite can be shown with small magnifications. If to escape this a large magnification is used, focussing difficulties enter, because of the deformation of the surface caused by the rupture itself. In the copper-plating method of study this difficulty does not arise.

719. Difficulty of Avoiding Pearlite in Crossing the Widmanstättian Structure.—This is illustrated in Fig. B of Plate 42. At the upper part of the micrograph rupture strikes S-65°-E, and here it follows the direction of the white ferrite laths. But after passing thus for half an inch on this magnification it resumes its main nearly vertical direction, cutting across the black and white stripes of pearlite and ferrite. To confine itself wholly to ferrite in crossing this lattice-work would clearly imply a great lengthening of its path.

720. Incompleteness of the Selection of Ferrite as the Path of Rupture.—These cases show how incomplete the selection of ferrite and the avoidance of pearlite by the path of rupture may be in these relatively high-carbon steels. Fig. D of Plate 42, showing a crack which brought about the failure of a 60-lb Pennsylvania Bessemer steel rail after 32 years' service,[1] teaches the same lesson. For ⅞ in. at its right-hand end, on this magnification, the crack runs along the ferrite network, but throughout the remainder of the field it runs indiscriminately through ferrite and pearlite. Even such preference as it shows for ferrite is referable in part to the slag enclosures in that ferrite. Professor William Campbell[2] and Dr. G. B. Waterhouse[1] indeed find that rupture passes across rails indiscriminately through ferrite and pearlite.

Dr. P. H. Dudley,[3] in considering four classes of rail steel, finds little if any preference of the path of rupture for either ferrite or pearlite except in the third class, in which rupture shows some preference for ferrite especially at low temperature between 0° and −40°C. (+32° and −40°F.). These types are:

1. low-carbon Bessemer steel, with carbon	0.35	phosphorus	0.10
2. medium carbon " " " "	0.56 to 0.65	"	0.06
3. " " " " " "	0.50	"	0.10
4. basic open-hearth " " "	0.62 to 0.75	"	0.02 to 0.04

[1] This long service under "moderately light equipment" was brought to an end on Jan. 13, 1910 in "freezing weather," by the use of heavy "Consolidation" engines weighing about 83 tons. The failure is typical of the split head class. J. P. Snow, private communication, May 11, 1910.

[2] Private communications, Mar. 19 and Apr. 1, 1915.

[3] Private communication, May 28, 1915.

The third class differs from the second and fourth in being richer in phosphorus, and from the first in being richer in carbon. Its ferrite must therefore be richer in phosphorus than that of any of the others. The phosphorus naturally occurs chiefly in the ferrite, and hence the ferrite of the third class is richer in phosphorus than that of the second and fourth because of the greater phosphorus content of the metal as a whole. Its ferrite is also richer in phosphorus than that of the first class, because of the higher carbon content of the third than of the first class. With this higher carbon content goes necessarily a proportionally smaller quantity of ferrite, through which to distribute the 0.10 phosphorus which the metal as a whole contains.

We have already inferred that it is the great ductility of the ferrite which tends to prevent the path of rupture from following it. In the present case it may well be that the high phosphorus content of the ferrite of the third class deprives it of much of this shielding ductility, and thus leads to its attracting rupture. This explanation tallies with Dr. Dudley's observation that this preference for ferrite is most marked at low temperatures, for it is at these low temperatures that the embrittling effect of phosphorus is generally believed to be most marked.

The striking thing is that the preference for ferrite in these cases should be so slight in spite of the slightness of the deformation which accompanies rupture in all of them, for it is exactly when the deformation is slight that the preference for ferrite is reported as at its greatest. One inclines to believe that the reported preference of rupture for ferrite is exaggerated.

721. Preference of Rupture for Slag Masses.—The fact that the slag masses naturally occur within the ferrite, and that their brittleness would naturally attract the path of rupture, may have given an exaggerated impression of the attractiveness of the ferrite itself. This may well be true of No. 9 of Table 27, and more especially of No. 14, in which the proportion of slag is extraordinary and grossly pathological, apparently forming a nearly continuous skeleton along the ferrite network.

The ease with which a fibrous fracture can be developed in wrought iron is proof of the preference of rupture for the parting between slag and metal, and thus of the weakening effect of slag. The path actually followed in the case of a fibrous fracture is so much longer than the trans-crystalline path as to show that the unit resistance along these slag and metal partings in this fibrous path is much less than the unit trans-crystalline resistance.

The attraction of the slag masses for rupture is likely to increase with the carbon content. In ductile wrought iron, which consists almost solely of ferrite, internal notches caused by the sharp edges of broken slag masses may heal very much more readily, by the squeezing in of the ductile ferrite, as shown in Fig. D of Plate 26, than they could if they occurred in the far less ductile pearlite.

The ductility of wrought iron even in the presence of as much as 2 per cent. of slag has been offered as evidence that slag is not a source of weakness. But the far greater ductility of even this wrought iron along than across the

direction of rolling disposes of this contention. In a ductile wrought iron rich in slag, this slag has been drawn into threads so placed that they do not interfere with its longitudinal ductility, as explained in §814, but they do interfere greatly with its transverse ductility and strength.

Again, even so eminent an investigator as Stead[1] seems to question the damage done by slag, in reporting a case in which rupture in steel, apparently of intermediate carbon content, seemed to avoid the slag masses and "passed along the borders of the pearlite areas." Such evidence must be scrutinized very carefully.

Indeed rupture, though following the slag masses, may not happen to pass through those visible in any given section, as is shown by Fig. 109. Suppose that the steel billet *FGHJKLM* is broken by rupture running lengthwise. Suppose that the little circles on the face *FGHJ* represent the ends of rods of slag, their continuation being sketched in broken lines. If rupture prefers the slag rods, it will take those which give it the shortest path through the metal, in this case the rods *B*, *C*, *D*, and *E*, but not *NN'* and *PP'*, the only ones which a section at *FGKL* happens to cut. Hence the fact that the only slag rods in this section are not among those included by rupture in its path ought not to suggest that rupture has no preference for slag rods. This will be considered further in connection with §732.

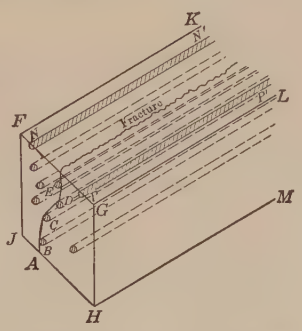

Fig. 109.—Rupture, though preferring slag bodies, may not pass through those seen in a given section.

722. Even Those Slag Bodies through which Rupture does not Pass may Weaken the Mass.—Following out the inference that rupture is likely to start in the less extensible pearlite rather than in the pliant ferrite, we see that the beginning and the end of the path of rupture are likely to be at spots determined by the breaking of the pearlite rather than by that of the slag bodies which are usually enclosed in the ferrite. In crossing the mass this crack will take the path of least resistance between its already existing beginning and end. It will naturally be only in a relatively small proportion of cases that this path will happen to pass in contact with a slag mass or through one. The slag masses which rupture thus avoids may not at first seem to be effective in weakening the steel as a whole. Yet if we look closer we incline to blame them. They lessen the cross-section of any given mass of ferrite in which they lie, and thereby lessen the resistance which it offers to the stress, and thereby in turn increase its tendency to stretch under that stress. This stretching, this failure of the ferrite to stand up under its share of the load, throws that share on the pearlite, and thereby exaggerates the overloading of the pearlite by the yielding of ferrite. These slag masses may thus indirectly be the cause even of rupture which does not pass through them.

[1] *Fifth Report Alloys Research Committee*, Excerpt, *Proc. Inst. Mechan. Engin.*, 1899, p. 80.

723. The Strengthening of the Ferrite during Deformation may Lessen Its Attraction for the Path of Rupture.—A most familiar effect of all forms of plastic deformation, whether by wire-drawing, hammering, bending, twisting, or otherwise, is great strengthening. We inferred in §715 that such attraction as ferrite has for the path of rupture is due to its weakness. The plastic deformation which precedes rupture may increase the strength of the ferrite very much more than that of the pearlite, and thus lessen greatly this attraction of ferrite.

Thus in the tensile test I find that the Brinell hardness of ferrite, in the form of the A.R.M. steel of 0.01 per cent. of carbon, increases by 106 per cent., whereas that of pearlite, in the form of steel of 0.92 per cent. of carbon, increases by only 10 per cent. The data are condensed in Table 25 (p. 466).

The Brinell hardness is roughly proportional to the strength. The fact that the deformation of the tensile test thus doubles the hardness of an isolated mass of ferrite, but does not increase that of an isolated mass of pearlite, helps us to explain why rupture has so slight a preference for ferrite in hypo-eutectoid steel, by permitting us to infer that in this intimate mixture of ferrite with pearlite a like greater strengthening of the ferrite than of the pearlite occurs during the progress of the tensile test, so that, by the time when rupture approaches, the initial relative weakness of the ferrite has been in large part effaced.

The rise of the stress strain diagram from the elastic limit to the maximum load reflects this strengthening. During this rise the diagram is approximately the locus of the elastic limit.[1] In the case of ferrite the degree of strengthening per unit of existing section is in greater proportion than this rise, because of the great reduction of area which occurs.[2] Indeed, this strengthening per unit of existing section probably continues during the remainder of the tensile test, from the maximum load till rupture. Clearly it is the strength of the existing section at the instance of rupture that influences the path of rupture.

724. Discussion.—To what extent this principle is applicable to the intertwined conglomerate of ferrite with pearlite of which hypo-eutectoid steel consists can be decided only by careful investigation. Attractive as the explanation is at first, on second thoughts it stirs our caution.

Though a mass of pearlite as a whole elongates less and hardens and strengthens less than a mass of ferrite, as shown in Table 25, yet because the pearlitic ferrite is identical in composition with the pro-eutectoid ferrite and differs from it only in being finer, it is conceivable that the very particles of this pearlitic ferrite between which rupture passes, those which rupture forces to release their hold on each other, might deform as much, and hence

[1] This is because, on releasing the stress and reloading the specimen at once, the stress needed to cause additional permanent stretch is very nearly that which existed at the moment of release.

[2] Compare Mesnager, *Internat. Assoc. for Testing Materials, Reunion des Membres Francais et Belges,* Mar. 23, 1907, p. 14.

strengthen as much before actually parting company, as the corresponding particles of pro-eutectoid ferrite, or even of the unimpeded ferrite of pure iron. This needs further study. Such evidence as I find tends indeed to show that the pearlitic ferrite does not become deformed to the degree to which the pro-eutectoid ferrite does. For instance, the pro-eutectoid ferrite of Fig. F of Plate 42 gives no indication of having elongated in rupture anything like as much as the ferrite of Fig. A of Plate 41,which exists in great masses forming practically the whole of the steel. If the ultimate yielding particles of the almost wholly unimpeded particles of ferrite of this nearly pure iron deforms so much more than those of the relatively impeded pro-eutectoid ferrite, that makes it easier to believe that this pro-eutectoid ferrite in turn deforms more than the far more impeded, indeed completely locked in, pearlitic ferrite.

Neither this evidence, nor the antecedent probability based on the reasons thus far given, really persuades us firmly that the deformation between those particles of the pearlitic ferrite which finally yield is less than that between those of the pro-eutectoid ferrite, and hence that the consequent strengthening of the pearlitic ferrite by the deformation is less than that of the pro-eutectoid ferrite.

Yet there is one good reason to hold that the pro-eutectoid ferrite strengthens relatively to the pearlite. Even if the strengthening by deformation is as great in the pearlitic as in the pro-eutectoid ferrite, that strengthening yet decreases the percentage difference in strength between pearlite and ferrite, because the excess of strength of the pearlite is probably due in part to the mechanical support which the pearlitic ferrite receives from the pearlitic cementite, and we see no reason why this fraction of its excess of strength should increase with the deformation.

725. Contrast between the Upheaval of Ferrite on Polished Surfaces and the Slightness of the Preference of Rupture for Ferrite.—This contrast is illustrated by Figs. E and F of Plate 27 on one hand and C of Plate 41 on the other hand. In the first two the pro-eutectoid ferrite network under pressure is squeezed out prominently from between the kernels of pearlite which it encloses, showing that it is much weaker than they. But in Fig. C of Plate 41, which represents the same specimen which meanwhile had undergone no additional treatment, the preference of the path of rupture for this same ferrite is very slight. This is referable to the principles already discussed, that though the relative weakness of ferrite in and by itself attracts the path of rupture, this weakness becomes relatively less during deformation, and that the pearlite itself, because of its brittleness, its inability to elongate without breaking, exerts strong counter-attraction for the path of rupture.

726. Why Deformation is Slight in Fatigue Testing.—At the end of a given pull, in fatigue testing, all the unexhausted elasticity of the neighboring metal which has not slipped tends to force slip in the ensuing push to retrace its steps along the very plane where it has just occurred, and thus to localize slip and keep it playing back and forth along a given plane. This localizing

of slip implies localizing of the deformation, so that the total deformation is only slight, as is shown by the fact that a surface polished before the fatigue test remains smooth, in great contrast with the roughening and upheaval which occur on a previously polished surface during the tensile test. In tensile testing, on the other hand, slip starting within a given grain meets resistance to propagation as soon as it comes to a terminal grain boundary. The existing and increasing stress may be relieved more readily by the starting of an additional slip along an additional plane in that same grain, without having to break down the opposition of the terminal boundary, which thus becomes strained elastically and moderately at many places where these successive slips approach it without breaking at any of them, though in the interior of each grain plastic deformation in the form of slip may have occurred at a great number of planes. Hence the deformation in the tensile test is widespread and cumulative, thus contrasting strongly with the relatively slight deformation of the fatigue test. In both tests rupture may be brought about by the exhaustion of the mobilizable crystalline metal, with the difference that in the fatigue test this exhaustion is localized along a small number of slipping planes, whereas in tensile testing it spreads out over a very great number.

Because the fatigue test is accompanied by relatively little deformation, it does not on one hand exhaust the little ductility of the pearlite masses, nor on the other hand does it strengthen the ferrite masses as the great deformation of the tensile test does. Thus it lacks both conditions which, in the deforming tensile test, tend to cause rupture to cross the pearlite, and hence it may be expected to develop in rupture a more decided preference for ferrite than is shown in the tensile test.

727. Why the Preference for Ferrite is Relatively Marked in Rupture Caused by Shock.—This is because shock causes relatively little deformation. But why does it? Let us consider the matter from the point of view of Beilby's amorphous theory. In tensile testing, as the metal slips in any part of any one plane and thereby throws an undue stress on the remainder of the metal in that plane, the increase of stress is so slow that, before the stress has grown strong enough to break through the grain boundary at the end of that plane, there has been time for the slipped metal to pass through the temporary weak mobile state and to reach the stronger amorphous state. This strengthening of the earliest slipping parts of any one plane along which slip occurs in itself opposes the continuation of slip along that plane, and thus coöperates with the disregistry at the end of the slip plane where it abuts against a terminal boundary to prevent the slip from propagating across the grain boundary. This change quickly brings the first slipping plane to a strength greater than that of other planes and as yet unslipped planes, either parallel to it or in general competing with it as paths of deformation, with the consequence that the next ensuing deformation occurs along first one and then another of these competing planes, and so accumulates and adds up to great total deformation.

But under shock the increase of stress is so rapid that it overloads the still unslipped residue of that first slipping plane more rapidly than the part which has first slipped gains in strength. Here is a race between the inconceivably rapid increase of stress caused by shock and the rapid, though apparently less rapid, strengthening of the contact along which slip has occurred, a strengthening representing the arrival of the metal at the stronger amorphous state after its passage through the weak mobile state. This rapid increase of stress causes slip, wherever it starts along any one plane, to propagate along that same plane, and thus to throw on the terminal boundary of the grain in which that slip has started, where that boundary is intersected by that slipping plane, a stress which is great enough not only to break across that boundary but to propagate this slip in like manner across and beyond the next grain, and so on. Slip started along a given plane thus propagates as a crack indefinitely, or with relatively few steppings across from plane to plane, essentially because the weakening of the total shearing strength along that plane due to the momentary slip at any one spot is followed up by an increase of stress more rapid than the increase of strength due to the setting of the metal made mobile by the first of the slip.

Even without the mobile-amorphous theory of slip, this same rapid increase of stress suffices to explain why shock leads to rupture with but little deformation, and hence strengthens the preference of rupture for passing through ferrite instead of pearlite, for the rise of the stress-strain diagram from the elastic limit to the maximum load, in spite of the simultaneous progressive loss of sectional area, proves that the elastic strength of the material does increase greatly during the deformation. Evidently it must be along the path where this deformation is occurring that this strengthening occurs. This strengthening alone can explain why shear, when it starts along a given plane, does not continue immediately completely across the test piece, so as to make the elastic limit also the maximum load.

We readily admit that this strengthening process, whatever its nature, may well occupy some time, even if only a little, so that a race immediately ensues between the tendency of deformation to propagate along its initial slipping plane on one hand, and on the other hand the strengthening of the contact metal along that plane by the deformation itself, so that it becomes strong enough to resist further deformation and thereby to throw it off to other and yet unstrengthened paths, with the additive effect of accumulating deformation along very many successive paths.

It is wholly in accordance with these ideas that the total elongation of a tensile test piece torn in two explosively may be very great.[1] This great elongation may be referred to incipient necking at numberless points along the test piece. The reason why necking may thus occur at very many points under explosive tension, though under quiescent tension it occurs habitually at only one point, is to be found in the frictional conditions in the necking section. In quiescent tensile testing the strain is so slow that but little

[1] The evidence of this will be given in another volume.

frictional resistance is offered by the necking section to its own outstretching by slipping along numberless slip planes. Hence once necking has started, the stress apart from the necking section decreases rapidly, as is shown by the familiar drop of the stress-strain diagram soon after passing the maximum load, simply because that necking section is unable to transmit great stress to the remainder of the test piece.

But in explosive tensile testing the frictional resistance in the necking section itself to forcible elongation is great enough to enable that necking section to transmit to the remainder of the test piece stress great enough to start that too stretching all along its length, and often to cause multiple necking. This tallies with the fact that after rupture by shock fine black lines may be found on a polished microsection a short distance back from the fracture. Shock causes these both in wrought iron and in steel, though much more prominently in the former. They cannot be seen before etching, which suggests to Rosenhain[1] that they are probably etching grooves, though possibly actual cleavage fissures, flowed over by the polishing film or invisibly narrow till widened by etching. Are they not Neumann bands? These are rarely caused by quiescent deformation (§584).

728. The Notched Bar Impact Test.—There are like reasons why the preference of rupture for the ferrite is relatively great in the impact testing of notched bars. First the dynamic conditions mean an almost instantaneous increase of the stress up to its maximum, in strong contrast with the long time available in the tensile test for the ferrite, amorphized along the slipping planes, to set, so that its elastic strength approaches that of pearlite. Thus the elastic strength of the ferrite does not increase during this impact test as it does in the tensile test.

Second, and probably more important, the great concentration of the stress at the apex of the notch tends strongly to start and propagate a crack, and thereby to avoid that serious local deformation which in the tensile test, according to our inference, strengthens the pro-eutectoid ferrite and thus favors the passage of the rupture through the pearlite.

729. The influence of a crack is evidently to concentrate the stresses on a minute section of the metal at its end, and thus to propagate itself through that section, causing but little deformation of the metal apart from the crack, because that very concentration of stress at the end of the crack shelters the metal elsewhere from stress. It is a matter of common observation that a sharp crack in propagating even through ductile metal creates but little deformation.

Though the effect of a crack in steel is like in kind to that of a notch in cloth or India rubber, it is much less in degree, as the following experiments show. Fine saw-cuts were made sometimes on one sometimes on both edges of steel and of wrought-iron test pieces; they were then closed at a temperature which though high was below the welding heat, thus practically

[1] *Journ. Iron and Steel Inst.*, 1906, No. II, vol. 70, p. 223.

making artificial cracks. These reduced the tensile strength of the remaining section as shown in Table 28.

A fine knife-cut on each edge of a strip of India rubber reduced the strength of the remaining section by from 60 to 70 per cent.

To arrest the development of a crack, Metcalf advises drilling a hole at its end.[1] A rounded notch or drilled hole increases the strength of a common tensile test piece per unit of remaining section, at the expense of the elongation.

730. Why a crack may have but moderate effect, as in the last case in Table 28, is explained by the dullness which its end often has, as in Fig. F of Plate 42, giving little suggestion as to the direction which it would probably follow if extended, in marked contrast with the terminal sharpness at *D* in Fig. C of Plate 41. Such rounded ends I have found usual in structures like that of Fig. F of Plate 42, composed of relatively coarse or at least coalesced bodies of ferrite.

TABLE 28.—EFFECT OF CRACKS ON THE TENSILE STRENGTH OF THE REMAINING SECTION. BAKER'S DATA[c]

	Wrought iron		Steel	
	Tensile strength, lb. per sq. in.	Loss, per cent.	Tensile strength, lb. per sq. in.	Loss, per cent.
Strength of solid piece	52,640	.....	72,800	.....
Piece cracked on one edge	.....	.....	55,328[a]; 37,856[b]	24[a]; 48[b]
Piece cracked on both edges	50,400	4.2	70,336	3.4

[a] When held in the usual way. [b] When held by a pin. [c] "The Working of Steel," *Proc. Inst. Civ. Eng.*, vol. 84, p. 164, 1886.

731. The Path of Rupture has a Strong Preference for Cementite Partings.—In marked contrast with the slightness of the preference of the path of rupture for the ferrite of hypo-eutectoid steel, even when that ferrite is concentrated in large masses, is the great preference which it shows for the contacts between masses of cementite, whether pro-eutectoid, primary or eutectic, and those pearlite or of eutectic.

It is this preference that causes the mirror fracture of hyper-eutectic cast iron, from which the name of the one variety, spiegeleisen, is derived. An extreme case is shown in Fig. B of Plate 40, the fracture of a great block of Müsen spiegeleisen or manganiferous cast iron. That the faces here exposed are those of cementite plates is shown by taking a section at right angles to them, on which these plates appear in cross-section as the great white parallel-sided areas in Figs. G, Plate 1, and N, Plate 3.

That this cementite is primary we have already seen in §196. The reason why the fracture is thus made up of large brilliant facets is that the adhesion between the smooth faces of the cementite plates and the eutectic which has later solidified in contact with them is so much weaker

[1] *Trans. Engin. Club, W. Penn.*, 1887, p. 133; *Journ. Iron and Steel Inst.*, 1887, No. II, p. 352.

than the cohesion either between the particles of the eutectic itself or between those of the cementite, that this face forms a natural parting so weak that rupture follows it accurately, neither leaving any eutectic adhering to the cementite nor tearing out any cementite particles to break the continuity of the surface and thus to dull it.

What is true of hyper-eutectic cast iron is true in a less striking degree of hyper-eutectoid steel, if its pro-eutectoid cementite is allowed by bad heat treatment to assemble in a network. In Fig. E of Plate 41, which shows the path of a crack across steel of 1.45 per cent. of carbon cooled slowly from 1,100° in order to assemble the cementite thus, rupture in general follows the parting between the cementite network and the pearlite kernels. Hence the invariable rule that hyper-eutectoid steel must not be allowed to cool slowly from above the transformation range, lest its cementite assemble in this pathogenic network, which is proof positive of bad heat treatment.

This explains also why the coarseness of the fracture of this steel increases with the length and height of the exposure to temperatures above the transformation range. With this exposure the grain size of the austenite increases, and with it in turn increases the coarseness of the pro-eutectoid cementite network which forms in slow cooling. This network, indeed, probably marks the outlines of those austenite grains.

In Hadfield's manganese steel, if through improper heat treatment it has plates of cementite, this tendency is even stronger, as is shown in Fig. E of Plate 42, probably because of the great contrast between the brittleness of the cementite and the extreme ductility of the austenite in which it is encased. Hence the rule that this steel must be heated highly enough to cause all its pro-eutectoid cementite to be reabsorbed in the austenite, in fact to 1,100°, and then cooled rapidly lest the cementite reprecipitate in cooling.

In coarsely crystalline silicon steel, also, Stead finds that rupture passes by preference along the very thin layer of cementite which sometimes lies between adjoining grains.[1]

732. Why the Preference of Rupture for the Cementite Partings of Hyper-eutectic Steel is so Much Greater than for the Ferrite Partings of Hypo-eutectoid Steel.—A sufficient reason lies close at hand in the relation which these two partings bear to the pearlite with which they are surrounded. That pearlite consists of about 6½ parts of ferrite to 1 of cementite. Hence a pro-eutectoid ferrite network is integrally united with about 86 per cent. of the substance of the pearlite by a ductile bond, the pearlitic ferrite, whereas a pro-eutectoid cementite network is attached to only about 16 per cent. of the substance of the pearlite, and this attachment is through narrow sheets of the brittle cementite itself, as shown in Fig. D of Plate 2. Note how thin these perfectly brittle sheets of the pearlitic cementite are, which alone bind the network cementite to the pearlite.

[1] Private communication, Mar. 24, 1915.

CHAPTER 30

FRACTURE

733. Introduction.—After some generalizations about fracture, this chapter takes up in §742 a review of the most important fracture studies ever made, those of Brinell, and in particular interprets them in the light of our present knowledge of the microstructure.

The conditions which lead to certain of these fractures are explained in Chapters 28, 29, and 31.

734. The chief types of fracture are:

1. *Granular*, or intergranular, caused by rupture passing along grain boundaries instead of across the crystalline grains themselves, Fig. A of Plate 40. The granular fracture is characteristic of steel and iron which have been heated above the solidus and are technically called "burnt," and in general of metals between the grains of which there is a brittle cement (§§685 *et seq.*).

2. *Crystalline*, trans-crystalline, or intra-crystalline, caused by rupture passing across the crystalline grains themselves.

Crystalline, as applied to fractures, has both a generic and a specific meaning. Probably all fractures which are not granular are trans-crystalline and in that generic sense crystalline. But the usual meaning of "crystalline fracture" excludes silky, hackly, fibrous, and columnar fractures, and refers to bright and usually brittle fractures, such as those along the crystalline cleavages or partings, as in Fig. G of Plate 15, and the martensitic ones. Such crystalline fractures occur in metals with either a strongly developed crystalline organization, such as phosphoric and overheated unhardened steel and bismuth, Fig. D of Plate 41, or with well-marked partings, for instance of cementite in overheated hyper-eutectoid steel, Fig. E of Plate 41, or in manganese steel reheated to 575°, Fig. E of Plate 42.

Among the crystalline fractures we may distinguish the cleavage ones, for instance, of brittle but unhardened steel, and the martensitic ones of hardened steel.

3. *Silky*, caused by the thread-like drawing out of the fragments of the crystals in the act of rupture, as in the case of very ductile metals like very low-carbon steel when fine-grained. The drawing out which causes this silky look is shown in Fig. A of Plate 41. The surfaces exposed in a silky fracture, for instance in the "cupped" tensile fracture of low-carbon steel, make an acute angle with the axis of the piece as if they were threads drawn out in rupture. Thus, to an eye placed nearly in a straight line between the fracture and the sun, the fracture looks its darkest when the axis of the piece is nearly or perhaps quite parallel to the light, and its brightest when that

axis makes a very obtuse angle with the light. Moreover the fracture evidently has a very great number of facets making all sorts of angles with each other but a nearly constant angle with the axis of strain, for when it has thus been brought to its brightest it does not change brightness when the test piece is rotated about its axis.

4. *Hackly* or hooked (German hackig), as of medium carbon steel when air cooled from above *A*3. The dullness and hackliness, like the silkiness of silky fractures, may be referred to the drawing out of the ferrite in the act of rupture. Each of the bright facets which we see in a hackly fracture is probably roughly plane, like the surface of a roughly planed wood. As we turn the specimen, successive lots of these specks flash out and again die away. It is true that each of these specks remains bright through a rather wide variation of the angle of incidence, for instance of direct sunlight, and thus gives the impression that it is rounded, for this is just what happens with the certainly curved surfaces of an inter-granular fracture, such as that of Fig. A of Plate 40. But this is wholly consistent with the rough planeness, for it is equally true of even a rather finely planed surface of metal.

5. *Fibrous* and *slaty*, as in the case of bars and plates respectively of wrought iron, Fig. C of Plate 40, of chrome and nickel steel and of nickel, Figs. G and I of Plate 42, and less prominently in those of most carbon steel.

6. *Columnar*, as in the case of the outer parts of ingots and other castings of quiet or "piping" steel, Figs. 15, p. 82, and H of Plate 18. The origin of this type is considered in §773.

Special forms of fracture which are in a sense crystalline are the dendritic,[1] shown in Fig. D of Plate 6, and those between masses of graphite or cementite and the eutectic pearlite, or other ground mass of the alloy, Fig. B of Plate 40.

All fractures of gray cast iron are graphitic, that is to say, the fracture passes through the graphitic masses and is coated with graphite, which can be removed readily with a wire brush. It is a matter of common observation, at least as regards high-carbon tool steel, that the fineness of the fractures is the finer the lower the finishing temperature, that at which the rolling or forging ends, and that its "fieriness," that is the presence of bright specks, represents unduly high heating. Hence four types of fracture tell four tales, thus:

Heating temperature	Right	Too High	Right	Too High
Finishing temperature	Right	Right	Too High[2]	Too High
Fracture	Dull	Fiery	Dull	Fiery
	Fine	Fine	Coarse	Coarse

[1] The fracture shown in Fig. D of Plate 6 is certainly crystalline in one sense, because it is made up of little pine-tree crystals, though it is wholly unlike what is usually called a crystalline fracture. It formed one side of a crevice which a steel casting tore into itself in cooling. Clearly this crevice must have opened during the later part of the passage down through the solidification range, when enough lower carbon metal had already solidified to give the mass as a whole a degree of solidity enabling it thus to pull itself apart instead of continuing to stretch like soft molasses candy, but when enough higher carbon metal still remained molten to assemble after rupture, under the crystalline force, into little pine-trees, yet not so much that it trickled down as it would have had there been much more of it. [2] *E.g.*, because the forging is stopped prematurely.

735. The internal surfaces of weakness which may be disclosed by these divers kinds of fracture may for convenience be classified thus:

Grain boundaries, disclosed in granular fractures.

Contacts between metal and the slag of wrought iron, slaty in the case of plates and fibrous in that of bars.

Contacts between different normal constituents, between pearlite and pro-eutectoid cementite in hyper-eutectoid steel, between the ledeburite eutectic and primary cementite in hyper-eutectic white cast iron.

Contacts between graphite plates and the enclosed metallic matrix.

Slip planes.

Cleavages.

Partings.

736. Order of Merit as Indicated by the Fracture.—Both for the hackly or pearlitic and the crystalline fractures, the finer the structure the better is the condition of the steel for most purposes.

The brittleness of the quenched state must usually be mitigated, "drawn" or "tempered" by reheating, which transforms the brittle martensite more or less fully first into troostite, then into sorbite, and then into granular pearlite. The temperature to which this reheating should be carried increases with the ductility and resistance to shock needed, at least up to about 400°. At this temperature the chemical transformation from the hardened to the unhardened state, from troostite to the sorbitic emulsion of ferrite and cementite, is probably complete. Further heating, which probably involves some coarsening of the fracture, may be needed in order to relieve the quenching stresses, and it is very often given in order to increase the ductility and shock-resistance. This it certainly does, but at the sacrifice of strength and elastic limit. There is much to favor the belief that, except for the relief of stress, any such heating far beyond 400° is unwise, and that the additional ductility sought in this way could be given better by lessening the carbon content. To vary the presentation, this sorbitic structure corresponds to the highest combination of strength with ductility and shock resistance for carbon steel of any given carbon content. If, with a steel of given composition, this structure does not give the shock resistance needed for the service in view, that resistance can probably be had to better advantage by substituting another steel of greater shock resistance while retaining the sorbitic state, than by retaining the present steel and replacing its sorbitic structure with another, provided always that so low a heating relieves the quenching stresses sufficiently for the service in view, which it often does not.

An exception may have to be made as regards the fibrous and leafy fractures, the *H* and *G* of Brinell. These correspond apparently to a special degree of ferrite banding, and this may have special merit for certain purposes.

The columnar and the granular fractures in finished products are pathological, and should not be tolerated under any usual conditions. Let us now consider certain of these fractures more fully.

737. Granular fractures[1] usually differ from crystalline ones in being curved instead of being made up of plane facets, and in being less brilliant though really smoother. They are surface-tension surfaces, as distinguished from the plane cleavage or parting surfaces which crystalline fractures represent. This accounts for their smoothness. Representing as they do the passage of rupture along the grain boundaries, they suggest if they do not imply the presence of some weak parting there, such as the local intergranular concentration of higher carbon and higher phosphorus matter through incipient fusion when medium and low-carbon steel are heated above the solidus, the austenite-cementite eutectic when very high-carbon steel is thus heated, and the pro-eutectoid cementite network when any hyper-eutectoid steel is cooled slowly from above the line *SE* of Fig. 23.

The "detail" fractures of boiler plates are sometimes intergranular, though even in these cases if an intergranular crack thus started is prolonged by a sudden shock "it invariably passes through the crystal cleavages and never along the boundaries."[2]

The weakness of the cementite partings, and their leading to a crystalline fracture, have already been considered in §§731, 732. The fracture which they cause is intergranular as regards the pearlite which breaks away from the cementite, but it might be called crystalline as regards the cementite itself.

738. Crystalline Fractures.—Typical crystalline fractures in Hadfield's silicon steel and in very slowly solidified low-carbon steel are shown in Figs. F and G of Plate 15. Stead's rectangular crystalline fracture of low-carbon sheet steel is shown in Fig. F of Plate 40. Let us first consider the cleavage fractures.

The brilliancy of the crystalline fracture of a given grain is due to its various parts being strictly parallel, so that all reflect their maximum of light simultaneously in one and the same position of the specimen, while its being less smooth than the granular fracture is due to its being made up thus of many parallel facets. Rupture travels along a given crystalline plane for a certain distance, and then steps across to an exactly parallel one.

This combination of properties in a crystalline fracture is seen by comparing Fig. G of Plate 15 and Figs. B, C and D of Plate 35, which all show the same fracture under different magnifications and illuminations. All the surfaces seen in Fig. G of Plate 15 are as brilliant as mirrors. The differences in shade are due solely to varying incidence of the main body of light, quite as parts of a brilliantly polished silver object look black, and the blacker the more perfect the polish.

[1] The causes of an intergranular path of rupture, leading to a granular fracture, are considered at length in Chapter 28, p. 489. The German "grobkörnig krystallinisch" used in the translation of Brinell's Swedish name for the coarse martensitic fracture of hardened steel, should be translated not "coarse granular crystalline" but "coarse crystalline." The literal translation "coarse-grained crystalline" would be unfortunate, because of its suggestion of "granular." These martensitic fractures are distinctly trans-crystalline and thus the opposite of the intergranular ones to which I advise restricting "granular."

[2] J. E. Stead, private communication, Mar. 24, 1915.

This mirror brilliancy suggests perfect smoothness. But greater magnification, as in Fig. B of Plate 35, shows that what looks in Fig. G of Plate 15 like a single mirror face is in fact made up of many parallel faces. The lines in Fig. B of Plate 35 are short steps from one facet to another parallel one, and the so-called acanthus-leaf figuring is like that in the cross-fracture of a slaty mass. That the fracture, in spite of its brilliancy, is not smooth is shown further by Figs. C and D of Plate 35 taken with oblique illumination.

The crystalline nature of this fracture is shown by the rectangularity of the masses cleft out,[1] in Figs. G of Plate 15 and C of Plate 35. In this latter the face photographed is a cubic cleavage, and the top and right-hand side are the two sides of the cube face. So with the top of the mass shown in Fig. D of Plate 35. So with the fracture of which the edge is shown in Fig. E of Plate 40. Note that here the fracture follows the cubic cleavage, as indicated by the direction of the cubic etching figures, both in the light and in the dark grain, in the latter indeed following first one and then the other of these directions. On looking closer we see that the longer part of this fracture in the dark grain is not strictly parallel to the etching figures, as if it stepped across at short intervals from one cleavage plane to another parallel one. So, too, with the rectangular mass of which three faces are shown after deep etching in Figs. E, F, and G, of Plate 35. The broken white lines indicate the edges of the several rectangular faces. That the Neumann bands of Figs. C and D of Plate 35 and the etching grooves of Figs. E, F and G are geometrically related to the sides of these rectangular faces has already been shown in §590. So with Fig. F of Plate 15, showing a piece of very crystalline silicon steel of about 4 per cent. of silicon, kindly given me by Mr. W. E. Ruder.

In Fig. B of Plate 35 the Neumann bands can be seen running diagonally. The acanthus markings seem to follow by preference either these same diagonal directions, or else a horizontal direction, parallel to the side of the cube face.[2]

What are often called "crystalline" fractures are probably only coarse hackly ones.

Of a radically distinct nature are the martensitic crystalline fractures of hardened steel, such as are discussed in 752 to §§755. Their nature remains to be determined. Their faces are not accurately plane, for each of their bright specks remains bright through a rather wide variation of the incidence of the light. Because the brittleness which accompanies these fractures prevents any drawing out of the particles in the act of rupture, any hackliness, one conjectures that each of these specks is a rounded surface, as if the martensite derived from each austenite grain yielded a conchoidal or glassy fracture, or as if rupture followed the austenite grain boundaries, as so many grains of

[1] The fragments marked A, B and C formed part of a thermit casting of nearly pure iron belonging to Professor William Campbell. In bending this casting these rectangular fractures formed.

[2] Compare with this case shown in Figs. B, C, and D of Plate 35 the cubic cleavage fracture of ferrite shown by Osmond, Fremont, and Cartaud, *Rev. de Metallurgie, Mem.*, 1904, vol. 1, p. 42.

quartz sand in a quartzite would, in spite of the perfection of outward crystalline form to which quartz so strongly inclines. Should this be confirmed we might divide the crystalline fractures of steel into the "cleavage" and the "conchoidal" ones.

739. Crystalline Fractures Accompanied by High Crystalline Organization.—The thorough organization of the crystalline structure which the strongly marked cubic cleavages suggest is naturally accompanied by very marked heterogeneousness of the physical properties. Thus Osmond noted that, though such crystalline iron cleaves readily and with but little deformation under the action of a chisel driven parallel to one of these cubic cleavages, it does not split but instead cuts if the chisel is turned 45°.[1] He believed that ferrite was peculiar in this respect, differing from the ductile metals in general on one hand, and from the generally brittle metals on the other. Austenite or gamma iron, as we know it in Hadfield's manganese steel and in 25 per cent. nickel steel, resembles copper and probably most of the other ductile metals in being ductile in every direction, that is to say plastic under all forms of deformation, whether intergranular or trans-crystalline. Bismuth, on the other hand, is brittle no matter to which of these classes its deformation belongs. But it remains to be shown whether this combination of brittleness along the cleavages with ductility in other directions is a peculiarity of ferrite, or only characteristic of the high degree of crystalline organization of Osmond's specimen, and whether other ductile metals, if highly enough crystallized, will not behave in like manner. That the organization of the specimen of Fig. F of Plate 15 is very high is shown by the ease with which Neumann bands are developed in it, even by quiescent pressure at the room temperature, though in less highly organized specimens they cannot be developed under these conditions (§§523, 830).

740. Stead's rectangular trans-crystalline fracture in ferrite, Fig. F of Plate 40. Stead found that a very coarse crystalline structure and great brittleness were induced in thin sheets of very low-carbon steel by rolling them at a relatively low temperature, below 700°, and then annealing them. Like effects have been noticed in some American works not only in sheets but also in bars. Here the evidence of crystalline structure is very clear. In those sheets in which this brittleness is extreme, the fracture runs in two directions at right angles not only to each other, but also to

[1] *Idem*, p. 45.

Note to Fig. 110.—This figure represents the conditions and results of experiments on steel from one and the same ingot, containing carbon, 0.52; silicon, 0.13; manganese, 0.48; phosphorus, 0.026; sulphur, trace. Each line beginning with *O* and ending with *V* represents one experiment. In each case the metal is gradually heated to a certain temperature, indicated by the point at which the line doubles and begins to descend. In most cases the temperature now descends without interruption; but in the last tests the cooling is interrupted, as indicated by a second doubling and the re-ascent of the line. *Stahl u. Eisen*, Nov., 1885, vol. 5, p. 611.

The slow cooling took place in dry "Kohlenstibbe" or charcoal dust. (J. A. Brinell, private communication, May 17, 1915.) This method was inapplicable in those cases in which the slow cooling was interrupted before it was complete, and was then succeeded by a reheating, or by a quenching. In these cases the slow cooling took place in the air. The specimens were $\frac{3}{8}$ in. thick.

For the condition of the carbon in Expts. 25, 26, 33, 34, 36, 41, 42, 48, 49, and 50, see §759.

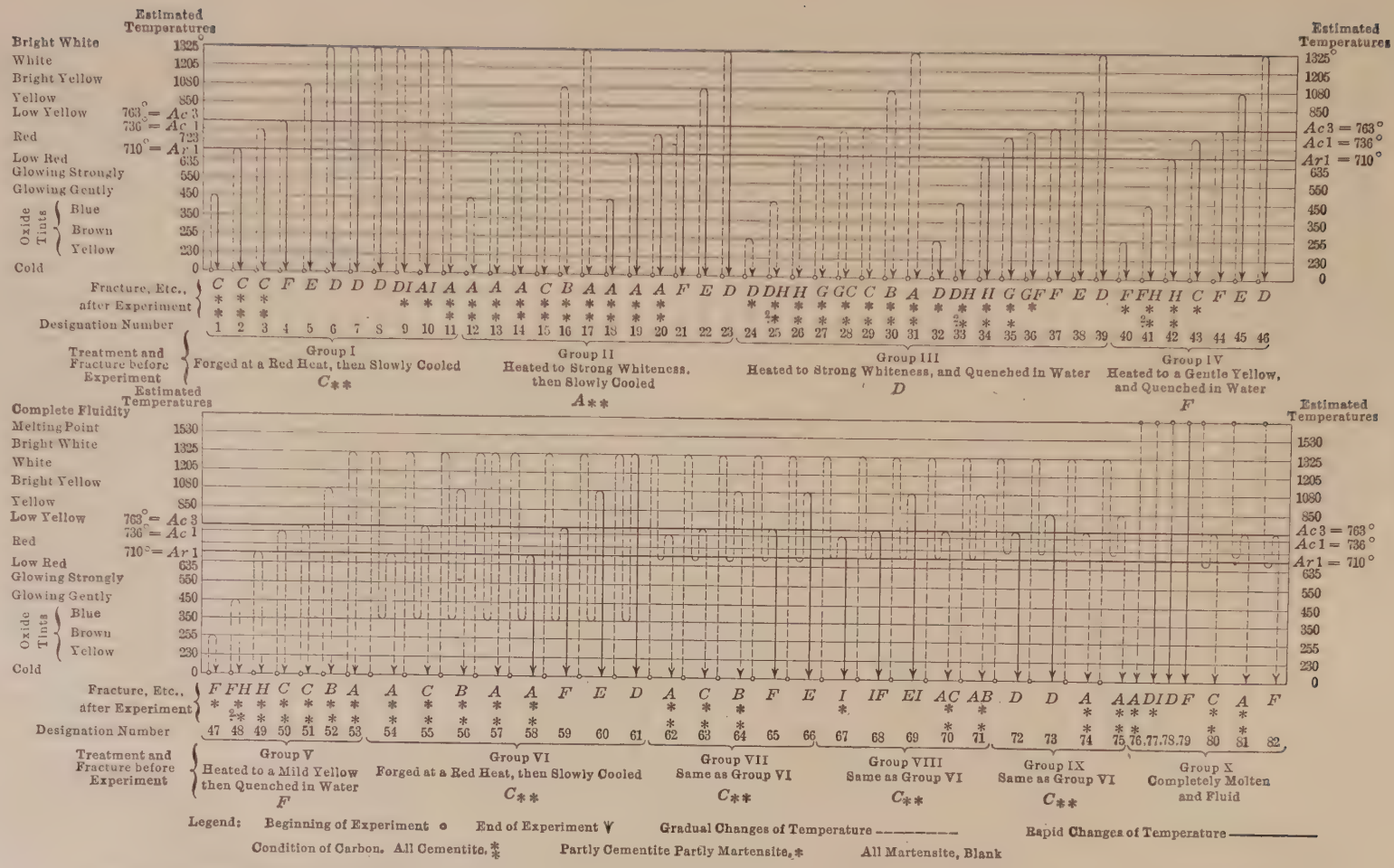
Estimated Temperatures
Bright White 1325°
White 1205
Bright Yellow 1080
Yellow 850
Low Yellow 763° = Ac 3
736° = Ac 1
Red 723
710° = Ar 1
Low Red 635
Glowing Strongly 550
Glowing Gently 450
Oxide Tints: Blue 350, Brown 255, Yellow 230
Cold 0
Fracture, Etc., after Experiment
Designation Number
Treatment and Fracture before Experiment
Group I Forged at a Red Heat, then Slowly Cooled C**
Group II Heated to Strong Whiteness, then Slowly Cooled A**
Group III Heated to Strong Whiteness, and Quenched in Water D
Group IV Heated to a Gentle Yellow, and Quenched in Water F
Complete Fluidity
Melting Point 1530
Group V Heated to a Mild Yellow then Quenched in Water F
Group VI Forged at a Red Heat, then Slowly Cooled C**
Group VII Same as Group VI C**
Group VIII Same as Group VI C**
Group IX Same as Group VI C**
Group X Completely Molten and Fluid
Legend: Beginning of Experiment o End of Experiment V Gradual Changes of Temperature Rapid Changes of Temperature
Condition of Carbon. All Cementite, ** Partly Cementite Partly Martensite, * All Martensite, Blank

FIG. 110.—Effect of heat treatment on fracture. Graphical representation of Brinell's results. Steel of 0.52 per cent. carbon.

the plane of the sheet itself,[1] so that the bodies cleft out by rupture are strictly rectangular and hence geometrical.

The strong contrast between the brittleness along these rupture planes and the ductility in other directions is shown by the degree through which the corners at shown in Fig. F of Plate 40 are dog's-eared without breaking.

This brittleness, like the coarsening of the grains of ferrite discovered by Sauveur, is brought about most readily by a certain critical degree of deformation, followed by annealing (§492). It will be discussed in another volume.

741. Slaty and Fibrous Structures.—Nickel and chrome-nickel steel plates often yield a fracture so like that of hickory wood as to suggest strongly that they contain slaty partings analogous to the cinder layers of wrought-iron plates. This fracture is especially prominent in the Krupp armor plate.

Indeed a fibrous fracture can usually be developed in carbon steel also, as will be shown in §749, and explained in Chapter 31. This capacity for developing a fibrous fracture results from some form of heterogeneousness.

742. Brinell's Fracture Studies.—The results of his sytematic study of the fracture of steel of 0.52 per cent. of carbon are set forth in his diagram, reproduced with slight amendments as Fig. 110, and condensed in Table 29. These give also my interpretation of the microstructures in which his various treatments result, and of the temperatures.[2] Each line, starting at a little circle on the horizontal axis and ending at an arrow head there, represents a single experiment. The fracture before the experiments is shown below the large braces, the resultant fracture immediately below the horizontal axis. Quenchings are indicated by unbroken lines, slower movements of temperature by broken lines. His slow coolings were only moderately slow, in charcoal powder when this could be used and in other cases in the air.

In those cases in which the inferred microstructure does not seem to correspond naturally to the observed fractures, I have had Brinell's experiments repeated, and have determined the microstructure, with temperature

[1] Stead proved this by showing that the fracture looked the brightest, and therefore reflected the greatest quantity of light to the eye, when the pencil of light made with the plane of the sheet an angle equal to that between that plane and a line from the fracture to the eye. *Journ. Iron and Steel Inst.*, 1898, No. I, vol. 53, p. 180.

[2] These estimates of the temperature have since been confirmed and in one case corrected by Dr. Brinell (Private communication, May 17, 1915). He estimates the temperature W as "perhaps 800°." My investigations lead me to put it at about 763°, the *Ac*3 of this steel.

Notes to Table 29.—[1] In two cases, Expts. 24 and 32, the steel, hardened by quenching from "strong whiteness" and therefore with the carbon all hardening, was reheated to about 255°. This of course converted a small part of the carbon into cementite, which would be in such fine emulsion that it could not be recognized. This slight reheating of the martensite does not change the microstructure visibly, and the present experiments show that it does not change the fracture.

[2] In all these cases the steel, hardened by quenching from "strong whiteness" or a "gentle yellow," was next reheated to 690°. This should certainly convert all its carbon into cementite. Brinell's test by nitric acid spotting indicated that the carbon was partly in the hardening state and partly as cementite. But here this test must mislead.

TABLE 29.—BRINELL'S STUDIES OF THE FRACTURE OF STEEL OF 0.52 PER CENT. OF CARBON. CONDENSED STATEMENT

Designation letter of fracture	Brinell's descriptions			No. of cases	Microstructure, observed or inferred		Treatment	Brinell designations	
	Description of fracture		Carbon condition, hardening (H) or cement (C)		Pearlite	Ferrite		Designation number of experiments	Designation of fracture
	Structure	Color							
A	**Coarse hackly. Facets shining**	Bluish	C	17	Sorbitic	Coarse network	Slow cooling after exposure to 1,300° without again heating to Ac3	11 to 14, 17 to 19, 31, 53–4, 57–8, 62, 75–6, 81	A
B	**Medium hackly. Facets shining**	Bluish	C	5	Sorbitic	Medium-sized network	Slow cooling after exposure to about 1,080°	16,30, 52, 56, 64	B
C	**Fine hackly. Facets shining**	Bluish	C	5	Sorbitic or fine granular	Sorbitic or fine granular	Sorbitic pearlite of fracture C heated to between 500° and about Ac1–5° and cooled slowly.	1, 2, 3	C
							Martensite heated to about Ac1–5° and cooled rapidly or slowly	43, 50	
				2	Sorbitic	Sorbitic	Martensite heated to about Ac3 and air cooled	29, 51	
				4	Sorbitic	Sorbitic	Coarsened network reheated to Ac3 and air cooled	15, 55, 63, 80	
D	**Coarse granular crystalline. Facets silver shining** (Metcalf, very lustrous)	White (Metcalf, yellowish)	H[1]	7	Long thin martensite needles from coarse austenite	Absent, its iron forming part of the martensite.	Quenched from about 1,300° without reheating beyond 300°.	6, 23, 24, 32, 39, 46, 61	D
				1			Quenched from about 1,080° after air cooling from about 1,300°	7	
				4			Quenched from about 760° or 850° after air cooling from about 1,300°, or from molten	8, 72, 73, 78	
E	**Finegranular crystalline. Facets silver shining** (Metcalf, fiery)	.....	H	6	Similar needles from intermediate sized austenite	" "	Quenched from about 1,080°, without immediately preceding exposure to a higher temperature	5, 22, 38, 45, 60, 66	E
F	**Bright porcelanic,** no decided crystallization visible to the naked eye	.....	H	5	Fine martensite from fine austenite	" "	Heated to about Ac3, quenched, and not again heated beyond 300°.	4, 37, 40, 44, 47	F
				4			Heated from below Ar1 to about Ac3 and quenched.	21, 59, 65, 82	
				1			Quenched from the molten	79	
H	**Porcelanic;** dull; fibrous. Wholly without crystallization	Dark or ash gray	C[2]	4	Granular; from reheating martensite to 690°	Banded	Martensite reheated to about 690°, and cooled either rapidly or slowly	26, 34, 42, 49	H
					The pearlite and ferrite may be in alternate bands				
G	**Leafy crystalline. Facets shining**	Bluish	H or C	2	Granular, banded, from reheating martensite to about 750°	Banded	Martensite reheated to about 750° (Ac3 − 10°), and cooled rapidly or slowly	27, 35	G
I	**Coarse crystalline. Facets shine like dull beaten silver**	.....	H	1	Martensite		Quenched from about 750° (Ac3 − 10°), after starting the realescence, by cooling to Ar1, 710°	67	I

determinations by means of Le Chatelier pyrometer as usual.[1] The rest of this chapter, except §748, is devoted to a study of his results.

743. The general division of his fractures is into those of annealed steel, *A*, *B*, *C*, *G*, and *H*; those of hardened steel, *D*, *E*, and *F*; and the transition fracture *I*. The true hardened fractures, *D*, *F*, and probably *E*, may occur also in steel which after hardening has been tempered slightly, by reheating to about 250°.

744. The fractures of unhardened steel are:

Hackly, his coarse *A*, intermediate *B*, and fine *C*;

Leafy crystalline, *G*; and

Dull porcelanic or so-called "*amorphous*," *H*.

All these occur in wholly unhardened steel, that is, in steel which is either fully annealed, or if it has been hardened, has later been heated at least to 690°, and has thus changed fully from the martensitic state which it had when hardened to ferrite plus cementite. The structure should be sorbitic or that of granular pearlite in all these cases, if, as I infer, the slow cooling was approximately of the order of rapidity of air cooling, for with specimens of this thickness this is not slow enough, according to my experiments, to give rise to lamellar pearlite.

In this medium carbon steel the coarse hackly fracture *A* is characteristic of that which has cooled slowly from a high temperature, say 1,325°; the intermediate hackly, *B*, of that which has cooled slowly from a moderate temperature somewhat above the transformation range, say 1,080°; and the fine hackly *C* of that which has been heated to or slightly above *Ac*3, and then cooled slowly. *C* occurs also in specimens the rolling of which is continued till the temperature sinks to about *Ar*3, even if their temperature is raised later to any point not above *Ac*3, as in Expts. 1, 2, 3, and 15. The refining of the coarse *A* to the fine *C* in this last experiment shows that *C* would have resulted from a like heating of steel initially yielding fracture *C*.

In brief, *A*, *B* and *C* respectively belong to such steel with its ferrite in a coarse network, in a network of intermediate size, and in fine masses, all three having their pearlite either sorbitic or granular because of the relatively rapid cooling.

745. The Degree of Coarseness of his Unhardened Fracture Corresponds to that of the Ferrite Masses.—That it is not affected by the structure of the pearlite is hardly surprising, because the nearly uniform rate of heating in all the experiments which do not cause hardening results in a nearly uniform structure for the pearlite. That the coarseness of the fracture corresponds to the size of the ferrite masses is shown clearly by Table 29. Note that the coarse *A* is always accompanied by a coarse ferrite network, the intermediate *B* by a network of intermediate size, and the fine *C* by sorbitic or finely granular ferrite.

[1] The steel used for these check experiments contained carbon 0.46, manganese 0.466, and phosphorus 0.041.

746. The Refining of the Coarse Hackly Fracture.—The correspondence which we have just noted between the coarseness of the ferrite masses and that of the hackly fracture is shown even better by a comparison of the cases which I collect in Fig. 111. In each of these a coarse ferrite network is first created by cooling slowly from about 1,300°. There are two common ways of destroying this network. The first is that of reabsorbing it by heating above *Ac*3, and then quenching in order to prevent it from reforming in cooling through the transformation range. The martensite thus formed is usually next converted into sorbite or pearlite by reheating to some temperature below *Ar*1, or even into fine-grained austenite by heating to above *Ac*1. This method does not apply to our present case.

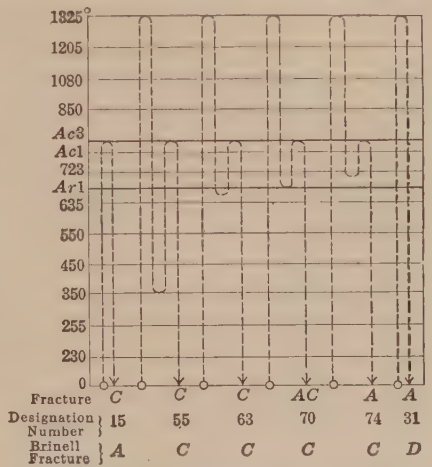

FIG. 111.—To refine the fracture from *A* to *C* requires refining the pearlite by heating above *Ac*1. From Brinell's data.

The other consists in refining the austenite by the double step of transforming it into pearlite by cooling past *Ar*1, and then back into fine austenite by heating past *Ac*1; and further in replacing the coarse ferrite network with ferrite in fine masses by continuing this heating past *Ac*3 so as to reabsorb the ferrite, and then cooling slowly. In the formation of a coarse network and its refinement by this latter method five steps occur. These are:

I. The austenite is coarsened by exposure to a high temperature, such as 1,300°.

II. In cooling slowly through the transformation range the various coarse grains of austenite surround themselves with the pro-eutectoid ferrite which they generate, and thus give rise to a coarse ferrite network.

III. On cooling past *Ar*1 this austenite transforms into pearlite, surrounded by the coarse ferrite network.

IV. On again rising past *Ac*1 this pearlite transforms into very fine-grained austenite, which remains fine-grained till the temperature again rises well above *Ac*3, the ferrite network being reabsorbed as the temperature rises from *Ac*1 to *Ac*3 through the transformation range.

V. Such fine-grained austenite, in again cooling from *Ar*3 to *Ar*1, arranges in extremely fine masses the pro-eutectoid ferrite to which it gives birth.

Thus steps I and II generate a coarse ferrite network, while III and IV break it up, and V regenerates the ferrite but leaves it in fine masses, so that III, IV, and V collectively transform the coarse ferrite network into fine masses of ferrite.

747. Applications of this Principle of Ferrite Refining.—Applying these principles to the cases collected in Fig. 111, we note that Expt. 15 has been preceded by steps I to III, so that it starts out with the coarse ferrite net-

work and fracture *A*. The experiment itself supplies steps IV and V, and thus completes the cycle, and transforms the initial coarse ferrite network which led to fracture *A* into the fine ferrite which leads to fracture *C*.

Expts. 55 and 63 start out with the ferrite in fine masses and fracture *C*. The first heating and cooling in each of them give steps I, II, and III, and thus create a coarse ferrite network which would lead to fracture *A*. But the second heating and cooling in each supply steps IV and V, which, together with III which has just occurred, lead to fine ferrite and fracture *C*.

Expt. 74 in its first heating and cooling gives rise to steps I and II, but because this cooling does not pass below *Ar*1 it does not give step III, which is absolutely essential to the breaking up of the coarse austenite. That is to say, coarse austenite cooled into the transformation range but not transformed into pearlite by cooling below that range, when next reheated above that range remains as coarse as before. In order to be converted into fine austenite it must be regenerated, that is first transformed into pearlite by exposure to a temperature below *Ar*1, and then back into austenite by again heating past *Ac*1. Hence though the coarse ferrite network formed in the first heating and cooling of this experiment is reabsorbed in the second heating because this reaches *Ac*3, yet the coarse austenite grains, remaining uneradicated, in the final cooling surround themselves with ferrite, and thus again give rise to a coarse ferrite network, so that this final cooling, instead of being step V as in Expts. 55 and 63, is only a repetition of step III. Hence the coarse fracture *A*.

I interpret Expt. 70 as carrying its first cooling into but not through *Ar*1, so that only the more advanced part of the metal took step III, changing from austenite into pearlite, but that the less advanced parts did not thus cool below *Ar*1, did not actually undergo the recalescence, *Ar*1, and hence remained coarse austenite. This conception that some parts of the specimen, even in the plane of the fracture, are less advanced than others is justified in general by the fact that the heating of the outer parts must needs outrun that of the axial parts, and here in particular by the fact that the recalescence, the transformation on cooling past *Ar*1, starts at certain nuclei as such undercooled transformations in general do, and propagates thence through the rest.

In the second heating and cooling these more advanced parts took steps IV and V and thus completed the cycle, generated fine ferrite, and gave fracture *C*; but the less advanced parts, because they had not taken step III, to change their coarse austenite into pearlite, could not take either step IV or V in this second heating and cooling. Here then the second cooling was in effect a repetition of step II, and by distributing the ferrite again in a coarse network led to the coarse fracture *A*. Hence the fracture is part *C* and part *A*.

In brief, in Expts. 55, 63, 70, and 74 the cooling from about 1,300° was interrupted and reversed by reheating to *Ac*3. In 55 and 63 this reversal substituted fine ferrite for a coarse ferrite network, because the cooling passed

below $Ar1$ and changed the coarse austenite to pearlite. This in the second heating transformed into fine austenite, which in turn generated fine ferrite in the second cooling. Here regeneration of the austenite occurred.

In Expt. 74 the cooling did not reach $Ar1$, the austenite was not transformed into pearlite, and hence no regeneration of the austenite could occur, so that the reversal had no effect, and the resultant fracture was A quite as in the companion experiment, 31, in which no reversal occurred.

In Expt. 70 the cooling passed $Ar1$ in part of the metal but not in the rest, so that the reversal refined the austenite and thus at last led to a fine ferrite and a fine fracture C in the former part, but in the latter part left the austenite coarse, and hence led to a coarse ferrite network and coarse fracture A.

748. The Combination of Coarseness with Hackliness.—That the influence of old austenite grain boundaries should persist as paths of low resistance has already been explained in §367. Nevertheless the ferrite particles so bridge the impurities concentrated in these boundaries that they draw out in rupture into little threads, and give the hooked or hackly texture, along the coarse surfaces which still mark pseudomorphically the positions of the grain boundaries of the now departed austenite.

749. The dull porcelanic fibrous fracture, H, "wholly without sign of crystallization," and giving an impression of being "amorphous through and through," is caused by reheating martensite, whether of coarse or of fine fracture, to about 690° so as to convert it into sorbite or granular pearlite. This same sorbitic state is given by heating even to lower temperatures. Thus heating even to a temperature as low as 450° gives a mixture of fracture H with the pre-existing martensitic fracture, coarse D or fine F as the case may be (Expts. 25, 33, 41, and 48).

The extreme fineness of fracture H is referable to the extremely fine state of subdivision which is the essential characteristic of sorbite. Its fibrousness is referable to the white ferrite banding, such as is shown in Plate 43, which this reheating brings about in many steels (§786).

This lengthwise banding of the ferrite not only offers itself as a natural cause of the fibrousness of the fracture, but accompanies the fibrous fracture. I found markedly fibrous fractures in specimens of this 0.46 carbon steel thus treated, quenched from 1,100° and then reheated to 650° or to 740°. The degree of fibrousness was very variable, probably depending on the exact conditions of rupture.

The very strongly fibrous fracture of Fig. I of Plate 42 occurs in a nickel steel which has undergone like treatment, quenching from above $A3$ followed by a sorbitizing reheating to about 580°.

750. The Leafy Crystalline Fracture, G.—This forms in two cases, Expts. 27 and 35, on carrying the reheating of martensitic or hardened steel to about $Ac1$ or 725°, a short stage beyond that which leads to the sorbitic fibrous fracture H. What the essential condition is which leads to this fracture I can hardly surmise, unless it is a certain specific degree of banding brought about by heating slightly beyond the temperature at which the

banding reaches the degree which lead to the fibrous fracture *H*. Fracture *G* seems to be independent of whether the steel is hardened or not, because whereas in Expt. 27 the steel was cooled slowly and hence unhardened, in Expt. 35 it was quenched and hence hardened if, as seems extremely probable, the quenching temperature was above *Ac*1. The slightly higher temperature reached in the companion Expts. 28 and 36, certainly should be above *Ac*1, and thus lead to hardening in the latter but not in the former, yet the fracture of each is in part *G*.

My preliminary experiments suggest that the leafiness is connected with this banding, and is like that of the leaves of an elliptic spring, or of the edges of the leaves of a pamphlet which has been rolled tightly for insertion into a mailing tube, or the edges of the cards in a pack slipped slightly out of square. Such leafiness I find strongly marked in the fracture of my 0.46 carbon steel which, after quenching from 1,100°, had been reheated to 650°, or even to 550°. But this is just the treatment which in other specimens leads to a fibrous fracture. Thus one gets the impression that "leafiness" is a variant of "fiber," somewhat as the structure due to distribution of the slag in wrought-iron plates is leafy when compared with the fiber which such slag causes in bars of the same metal. Thus the leafy fracture shows rather the edges of successive leaves than the bodies of individual leaves.

751. Composite Fractures which are in Part G, Expts. 28 and 36.—These may be explained tentatively by assuming that certain parts of the specimen are somewhat more advanced than the rest, and that the former have reached *Ac*3 and thus lost the last traces of the coarse structure formed in the prior high heating which led to the initial coarse martensitic fracture *D*, while the latter have not reached *Ac*3, and hence retain this coarse structure. The less advanced parts accordingly have the leafy fracture *G* which after the slightly lower heating of Expts. 27 and 35 forms the whole, while the more advanced parts have, after the slow cooling of Expts. 27, the fine hackly *C* as in Expt. 15, but after the quenching of Expt. 36 the fine martensitic *F*, as in Expt. 21.

752. The Martensitic Crystalline Fractures of Hardened Steel, Coarse D, Intermediate E, and Porcelanic F.—These fractures represent in every case martensite resulting from quenching austenite of a degree of coarseness corresponding to that of the fracture itself, though this martensite may be accompanied by much troostite. The coarse *D* forms in these experiments when the austenite has been greatly coarsened by heating to about 1,300°, the intermediate *E* when it has been coarsened to an intermediate degree by heating to about 1,080°, and the very fine *F* when it has not been seriously coarsened, but is in the relatively fine state which it has when the steel is heated barely past *Ac*3.

The structure of the martensite itself, in this medium carbon steel, is affected relatively little by either the quenching temperature or the temperature to which the heating for quenching has been carried, two tem-

peratures which do not always coincide. Indeed, in these experiments of Brinell's, as I interpret them, the coarseness of the martensitic fracture follows that of the austenite grains, and is independent, as far as the record shows, of any changes in the structure of the martensite other than those due directly to the structure of the austenite whence it springs.

The long exposure to a high temperature which leads to *D* provides the great mobility which permits not only the coarsening of the austenite grains but also thorough crystalline organization, including the development of planes of marked weakness, along which rupture passes by preference. Even the intermediate heating to 1,080°, though it leads to only an intermediate grain size, yet seems to cause a sufficient degree of crystalline organization to cause planes of marked weakness, along which rupture passes. This tallies with the marked martensitic structure which compression develops on the previously polished surface of such steel as this, as shown in Fig. F of Plate 28.

The gentle heating to just above *Ac*3 gives little opportunity for growth of the grains of the new-born austenite, so that the structure remains extremely fine. To this fineness of grain the fine porcelanic fracture *F* naturally corresponds.

The brightness of all three fractures reflects the brittleness of the martensite, its inability to draw out during the act of rupture.

Let us now consider these fractures in more detail.

753. Coarse Crystalline Martensitic Fracture.—This occurs not only in all the cases in which the steel was heated to about 1,300° and then quenched immediately, but also in five in which, after exposure to such a temperature, it was cooled in the air to an intermediate temperature not below *Ar*1 and then quenched. These are Expts. 7, 8, 72, 73, and 78, which collectively represent such intermediate air coolings to about 1,080°, 850°, 763°, or *Ac*3, and about 715°, or toward the lower part of the transformation range. In Expts. 72 and 73, after the cooling to about 715°, the temperature was raised slightly before quenching, but this reversal can have had no important effect.

Here the austenite was coarse because of the high heating, but the resulting martensite might have been expected to be fine because of the low quenching temperature. In direct experiments I find that such incomplete coolings toward or even to somewhat below *Ar*3 have little if any effect on the structure of the martensite of this carbon content. Of course, when we heat directly to a high temperature and quench, the coarseness of the austenite corresponds to the quenching temperature, but it does not in these present cases in which a high heating is followed by quenching from a much lower temperature after an intermediate cooling.

When coarse austenite is made to give birth to a coarse ferrite network by cooling into the transformation range, as in Expts. 72 and 73, and is again heated to *Ac*3 so as to reabsorb that network, but without having passed below *Ar*1 so as to regenerate the austenite itself, and is then quenched,

the fracture is the same, *D*, as it would have been in case this network had not been thus generated and reabsorbed. Hence we infer that the austenite, into which this ferrite network transforms in rising through the transformation range, identifies itself fully with the austenite kernels which have not transformed during this dip into that range, so fully indeed as to leave in the fracture no trace of that generation of ferrite and its return to the austenitic state.

When such a coarse ferrite network is generated within coarse austenite by cooling into the transformation range, as in Expts. 9 and 77, and the steel is then quenched, thus converting the accompanying austenite into martensite, the presence of the ferrite network does not appear to affect the martensitic fracture, which remains coarse, *D*, quite as it it would have been had the ferrite been lacking. The reason why part of each of these two composite fractures is *I* will be considered under *I*.

754. E, the fine crystalline martensitic fracture, in every case represents martensite resulting from quenching a fine-grained austenite which had been formed by heating from below the transformation range to a moderate temperature, about 1,080°, and quenching (Expts. 5, 22, 38, 45, 60, 66).

755. F, the porcelanic martensitic fracture, every case represents martensite or troostite resulting from quenching fine-grained austenite when at or just above *Ac*3. In Expts. 40 and 47 the quenched steel was reheated to about 255°. This latter treatment, though it allows the transformation from austenite into ferrite plus cementite, arrested by the rapid cooling, to proceed to a certain degree, and thus softens and disembrittles the steel materially, yet has little effect on either its fracture or its microstructure, which remain pseudomorphically martensitic.

The martensite was formed in nine cases by heating to *Ac*3 and quenching (Expts. 4, 21, 37, 40, 44, 47, 59, 65, and 82); and in one case by quenching a very thin stream of molten steel in water (Expt. 79).[1]

The extreme fineness of this porcelanic fracture represents the fineness of the austenite, and its brightness represents the brittleness of the hardened state, the inability of the steel to elongate. The high quenching temperature above the melting point of Expt. 79 might have been expected by some to lead to a coarse-grained martensite, and hence to the coarse martensitic fracture *D*. Its fineness is an additional illustration of the principle pointed out in §753, that the coarseness of the fracture represents not the quenching temperature as such but the coarseness of the austenite whence the martensite forms. In this case the steel passed through the range of temperature above *Ar*1 so rapidly that the austenite grains had no opportunity to coarsen.

756. The Gradual Development of Fracture F.[2]—Starting with the coarse hackly fracture *A*, there is a progressive refining of the fracture as,

[1] I have to thank Mr. A. S. Hummell, Chief Chemist of the Taylor-Wharton Iron and Steel Company, for kindly preparing a specimen to represent Expt. 79.

[2] The Author, *Proc. Amer. Soc. Testing Materials*, 1911, vol. 11, p. 382.

with the rise of temperature through the transformation range from $Ac1$ to $Ac3$, the coarse ferrite network is reabsorbed by the surrounding austenite. This may be shown by heating a series of specimens which are initially of overheated medium carbon steel with coarse hackly fracture A, to a series of temperatures below, in, and above the transformation range and quenching, or by heating a single bar differentially and quenching it. Under these conditions the refining of the coarse A begins on rising past $Ac1$, with the beginning of the acquisition of the hardening power and of the reabsorption of the ferrite network.

The completion of the refining of the fracture coincides approximately with that of the reabsorption of the pro-eutectoid ferrite.

757. The Dull Coarse Crystalline Fracture I.—In Expt. 67, the only case in which the whole of the fracture is of this rare[1] type, the transformation from austenite to ferrite and cementite has evidently occurred, because if the first cooling had not started this transformation the fracture would have been D, as in Expt. 73. This transformation, having been started by that cooling, would have time to propagate across the whole cross-section during the following rise. Had this cooling been complete, as in Expt. 11, the fracture would have been coarse hackly, A, and so it would if the first cooling, instead of reaching $Ar1$ and starting the $Ar1$ transformation, had been to a somewhat higher temperature, as in Expt. 74.

The difference between Expts. 11 and 74 on one hand which yield A, and 67 which yields I, is that in the first two the $Ar1$ transformation has occurred in the usual way through continuous slow cooling, whereas in the last, after having been incited by touching $Ar1$, it has then completed itself at higher temperatures. This, I think, gives us the clue. In a common slow cooling the $Ar1$ transformation is accelerated by the fact that the temperature at which it occurs, $Ar1$, is well below the equilibrium temperature $Ae1$. Thus it occurs with undercooling, and therefore rapidly, and because of the undercooling generates lamellar pearlite. I suggest that the rapidity, or if you prefer the violence of the reaction, suffices to break up the prior austenite structure, which in the absence of that violence persists pseudomorphically through the transformation from austenite into ferrite and cementite.

This occurrence of the $Ar1$ transformation without the usual degree of undercooling may be true of every experiment in which fracture I results. In all of them the slow cooling, after touching $Ar1$ and thus starting the transformation, has been replaced either by a quenching as in Expts. 9 and 77, or by a rise past $Ac1$, as in Expts. 67, 68, and 69. Though in this heating the change from austenite to pearlite started at $Ar1$ would have time to complete itself before $Ae1$ was reached, yet its progress would be less rapid than under the usual conditions because of the slightness of the existing undercooling below $Ae1$.

[1] Even Brinell, probably the greatest authority on fractures, had rarely if ever seen it before these experiments (*Stahl und Eisen*, 1885, vol. 5, p. 618).

In those cases in which only part of the fracture is *I* we have to assume a more and a less advanced part of the specimen. Further assumptions that we need are that in 9 and 77 the more advanced part has barely transformed into pearlite and hence yielded fracture *I*, and that in 10 and 69 this more advanced part has transformed fully, resulting in *A* or *E* according to the further treatment, while the less advanced has not cooled well past *Ar*1 and hence yielded fracture *I*.

The reason why I refer this fracture to this supposed incomplete effacement of the coarse austenitic structure is that I find nothing else which these various specimens have in common. The *I* part of 9 and 77, for instance, must be in part at least pearlitic, that is it must have transformed at least in part from austenite into pearlite, for otherwise the whole fracture would have been *D*. 69, on the other hand, is most certainly all martensite. Thus *I* seems to occur both in unhardened and in hardened parts.

758. Composite Fractures.—In 13 cases the fracture combines two types. Five of these, Expts. 9, 10, 68, 69, and 77, we have already considered under fracture *I*, and two others, 28 and 36, under fracture *G*.

In four cases, Expts. 25, 33, 41, and 48, a martensitic specimen is reheated to about 450° and again cooled, with the result that its fracture is now in part martensitic and in part sorbitic, the dull procelanic *H*. On heating to 450° the transformation from martensite to sorbite ought to be complete. But yet the structural planes of weakness of the austenite may still remain in certain parts of the specimen. These parts would therefore yield the martensitic fracture *D*, which the whole would have yielded before this reheating to 450°. But in other parts of the specimen these structural planes may have been effaced so nearly completely that they no longer affect the fracture, which is therefore sorbitic, *H*, here.

In two cases, Expts. 70 and 71, the temperature is first raised to about 1,300° so as to coarsen the austenite, and then lowered to but apparently not through *Ar*1, and then raised again. Because of the habit of this transformation of propagating from nuclei, so that shortly after its start these are more advanced parts of the specimen which have transformed and less advanced ones which the propagation has not yet reached, in these experiments, when the temperature, having touched *Ar*1, was again raised, the more advanced parts had transformed into pearlite while the less advanced ones were still coarse austenite.

In Expt. 70 the temperature was next raised to or barely past *Ac*3. This would have no effect on the less advanced and austenite parts, but it would cause the more advanced and pearlite parts to change back into fine-grained austenite, with the result that the specimen would now consist of austenite partly coarse because unregenerated, and partly fine, because thus regenerated.

In the slow cooling which now followed, the coarse austenite would generate a coarse ferrite network, which would lead to the coarse fracture *A*, whereas the fine austenite would yield ferrite in fine masses, and this

would lead to fracture *C*. Hence the composite fracture *AC* actually found.

In Expt. 71 a like cooling into but not past *Ar*1 was followed by heating to about 1,080°, and then by slow cooling. By like reasoning the less advanced parts of the specimen, which had not transformed into pearlite on nearing *Ar*1, would remain coarse austenite, and in the final cooling would generate a coarse ferrite network and hence yield fracture *A*; while the more advanced parts which did transform into pearlite at *Ar*1 would generate fine austenite on rising past *Ac*1. This on reheating to 1,080° would coarsen to the intermediate degree corresponding to this temperature. This austenite in the final cooling would generate a ferrite network of like intermediate coarseness, and thus lead to fracture *B*. Hence the composite fracture actually found, *AB*.

759. Cement and Hardening Carbon.—Brinell determined the condition of the carbon by means of spotting with dilute nitric acid, of 1.23 sp. gr. Under these conditions the martensite of hardened steel yields a black-brown, sooty, amorphous layer of carbon, whereas the ferrite and pearlite of unhardened steel yield a bluish-black layer which gives a grayish-black streak and appears to be crystalline. The carbon of the cementite of unhardened steel is often called "cement carbon," whereas that of the martensite of hardened steel is often called "hardening carbon."

This test, which I have found to give clear indications under most conditions, seems to fail when the pearlite is very finely divided or sorbitic. Thus he gives the carbon as only partly cement in Expts. 26, 34, 42, and 49 in which the hardened steel was heated to *Ar*1, say 710°, and cooled either slowly or rapidly, and also in Expts. 25, 33, 41, and 48, in which it had been reheated to about 450°. In the former of these groups it is certain and in the latter it is probable that the transformation from martensite to sorbite or pearlite had completed itself, so that there was no martensite and hence no hardening carbon left. Moreover, in Expt. 36 the steel seems to have been heated very nearly to *Ac*3, and hence presumably above *Ac*1, so that the whole of the pearlite must have changed to austenite, and then to martensite on the quenching which followed. Under these conditions the carbon should be all hardening; yet he reports it as partly cement. In all these cases I have taken the liberty of correcting what I take to be an error due to the use of this spotting method.

CHAPTER 31

GHOSTS AND THE OTHER ELEMENTS OF FIBER

760. Summary.—After defining fiber, and showing the fibrous structure on various magnifications in §762, I show that though fiber is due to longitudinal heterogeneousness caused in rolling and forging, yet the tendency to follow the longer and fibrous path of rupture is increased by increasing the trans-crystalline strength (§764). The ultimate structural cause of fiber in wrought iron and plastic-origin steel is slag, but in molten-origin steel it is the dendritic segregation, which assembles carbon, phosphorus, slag, and blowholes, and to a smaller degree manganese, in the fillings between the dendrites of the initial solidification, whence they are drawn out lengthwise by rolling and forging.

In spite of its leading to fiber, the dendritic type of solidification is to be preferred to the onion type, because of the serious upper axial segregation to which the latter leads (§774). But the dendritic form is least harmful when it is pushed to the extreme so that the dendrites are closely packed, because then the local enrichments and irregularities in composition are not only less in intensity but less also in their absolute width, and hence more readily effaced by diffusion and mechanical kneading (§774).

The conditions which should lead to this close packing of dendrites are quiet and rapid solidification, the former induced by thorough deoxidation, the latter by low casting temperature, the use of thick-walled, cold, and narrow moulds, and by early stripping (§775).

The phosphorus banding in turn causes ferrite ghosts. The fiber in steel almost free from phosphorus is so marked that it must be referred to the dendritic segregation of some other element, perhaps manganese (§769). The life history of ghosts is given in §§778 to 786, and its explanation in §§787 to 797. Because slag does not diffuse at all and phosphorus only extremely slowly, fiber is extremely presistent (§796).

761. Introduction.—Fiber implies some longitudinal heterogeneousness. Wood is fibrous because it is stronger lengthwise than crosswise, and this in turn because it has certain internal longitudinal surfaces of weakness. When bent greatly it tends to yield by longitudinal shear, that is by the shearing or sliding of one fiber over another. This excess of longitudinal over transverse strength of the mass as a whole I take to be the essence of fiber. It has two aspects, the excess of longitudinal strength and the deficit of transverse strength.[1] Of these the former is usually the cause of

[1] We cannot insist that, as regards the ultimate particles, this difference is chiefly in strength rather than in ductility or flexibility; but as regards the mass as a whole it may be treated as a difference in strength.

the merit of fiber. We value a fibrous wood not because of its relatively small transverse strength but because of its absolutely great longitudinal strength.

There are indeed certain cases for which the deficit of transverse strength is in itself a merit. Were a rope as strong crosswise as lengthwise it would lose that flexibility which makes it useful for purposes for which an equally strong wooden or metal rod would be uselessly stiff.

In the absence of any agreement on the subject I propose to call all systematic differences between the longitudinal and the transverse structure and properties "fiber," whether these differences are properly regarded as giving longitudinal strength or transverse weakness or brittleness. All excess of longitudinal over transverse strength, and all tendency, whether developed or dormant, to break with a longitudinal or fibrous fracture, may be referred to fiber in this sense of the word.

762. Illustrations of the Fibrous Structure.—This fibrous or dendritic structure is shown in its natural size in Fig. C of Plate 15 and D of Plate 43. The light streaks in the latter are what are called ferrite "ghosts." These as developed by air and furnace cooling respectively in a steel of 0.147 per cent. of carbon are shown in Figs. A and B of Plate 43; as developed in the slow cooling of a very phosphoric steel in Fig. C; as developed by reheating quenched steels of 0.46 and 0.40 per cent. of carbon in Figs. E, F, and G of that plate; as developed in the common industrial slow cooling after rolling or forging in the transverse section of a $\frac{3}{8}$-in. "universal" boiler plate of 0.20 of carbon in Fig. E of Plate 44, in the longitudinal section of a $\frac{3}{8}$-in. "sheared" boiler plate of 0.20 per cent. of carbon in Fig. F, and in a hammered $\frac{1}{2}$-in. steel rod of 0.43 per cent. of carbon in Fig. G of that plate. In this last case they enclose slag rods.

Figs. E and G of Plate 43 show that these ghosts may be more prominent on small than on large magnification. Additional evidence of this is given by the greater prominence of the banding in Fig. E if seen from a distance of perhaps 50 ft. than if seen from nearby in the usual way. Ferrite ghosts formed in a special way by the graphitization of the cementite bands in a steel of 1.45 per cent. of carbon are seen in Fig. G of Plate 7.

Other manifestations of this fiber are the fibrous fracture of metallic nickel and of nickel steel shown in Figs. G and I of Plate 42, and of barked wrought iron in Fig. C of Plate 40. The conditions leading in Brinell's experiments to a fibrous fracture are given in §749.

In considering fiber in metal we should ask whether flexibility is likely to be useful, and whether fiber in any given case represents unusual longitudinal strength, and hence is a merit, or defective transverse strength, and hence is a defect.

763. The Cause of Fiber in Iron and Steel.—The longitudinal heterogeneousness which causes fiber is such that it leads the longitudinal or fibrous path to compete with the usual shorter 60° trans-crystalline path in determining rupture. The trans-crystalline path is not only much the

shorter, but also usually conforms much the more closely to the direction of the chief stresses. If, in spite of this, rupture is to be diverted to the fibrous path, the means at hand are to weaken this path, to strengthen the trans-crystalline path, and to adapt the stresses specially to this end.

The grains of hot-rolled wrought iron, for instance, are themselves equiaxed, but are separated into quasi fibers by layers of slag, like a mass of minute cubes of iron very highly magnetized, divided up into rows by thin strips of glass, the strength and ductility of the whole being due to the magnetization of the iron cubes, and being merely lessened by the glass. Such iron may be likened to a gneiss, the crystals of feldspar and quartz with their axes in all azimuths, the plates of mica lying parallel and forming partings. In this view wrought iron has fiber; but the fiber as such should weaken the mass.

764. Fibrous Wrought Iron Differs from Crystalline not in Containing More Slag but in being More Ductile.—Thus as between a given specimen of fibrous wrought iron and a crystalline one, there may be no difference as regards the kind, quantity, or distribution of the slag, but only a difference in the strength and ductility of the metal. One breaks with a fibrous fracture because the grains of metal are so strong that the path of least resistance, instead of crossing them, is mainly along the contacts of metal and slag, here and there stepping across from one layer of slag to another and in this cross stepping necessarily passing across the grains. The other breaks with a crystalline fracture simply because its metallic grains are so weak that the shorter path across them offers the least resistance. They are not strong enough to divert rupture to shearing longitudinally along the slag and metal contacts.

Much in the same way burnt iron breaks with a granular fracture because the burning has so weakened the intergranular contacts, the grain boundaries, that they form a path which is of less resistance than the competing paths, either the trans-crystalline ones across the grains themselves or the fibrous ones along the contacts of slag and metal.

765. Fiber-generating Longitudinal Defects.—There are three prominent sets of internal defects running lengthwise of rolled and forged pieces, and hence causing transverse weakness or brittleness and hence in turn favoring a fibrous or slaty fracture.[1] These are: (1) "sonims," that is, solid non-metallic impurities, including slag, oxides, sulphides, etc.; (2) unwelded blowholes; and (3) the ghost structure, called "ferrite banding," "ferrite ghosts," "micro-ghosts," etc.

All three classes represent the fillings *ab* of Fig. 112, p. 552, between the trunks *gd* and *ef* of the dendrites formed in the original solidification. Though these fillings are initially transverse to the ingot, yet they, or more likely their fragments, are usually drawn out into longitudinal rods or

[1] Rosenhain, *Internat. Assoc. Testing Materials*, Vth Congress, Copenhagen, 1909, I, 4; VIth Congress, New York, 1912, First Section, II, 2. Hibbard, *ibid*, II, 10; *Trans. Amer. Inst. Min. Eng.*, 1910, vol. 41, p. 803.

threads if the reduction is equal along both transverse axes of the ingot, as in rolling rods and rails, and into sheets if it is chiefly along one of these axes,[1] as in rolling plates.

It might be thought that the drawing out of the individual polyhedral grains of ferrite in cold rolling and like operations, shown for instance in Fig. I of Plate 44, would constitute a sort of fiber, but the evidence given in §821 indicates that it does not.

766. Slag is always and necessarily present abundantly in wrought iron, and is generally present to an appreciable degree in molten origin steel.[2] Slag enclosures in wrought iron are shown in Figs. F of Plate 14 and D of Plate 26, and in steel in F of Plate 43 and G of Plate 44. The shape is of the same general family in both.

Like the other substances present, slag tends to spheroidize by surface tension on long heating, though this process is incomparably slower than the equiaxing of the grains of ferrite drawn out by rolling. Sorby found that the slag in wrought iron which had been long heated was in almost perfect spheres.[3]

767. Blowholes are usually present in Bessemer and open-hearth steel, and especially in low-carbon steel, though they can be prevented even here by quieting or deoxidizing additions, such as silicon, aluminum, and titanium. It is possible to weld them up[4] and so to efface them in low-carbon steel, unless they lie so near the surface that their sides become coated with oxide.

768. The ghost structure,[5] shown in Plates 43 and 44 and diagrammatically in Figs. 113 and 114, consists of an alternation of light and dark bands, or of light bands running through a dark ground mass. The light bands

[1] Tyndall found that this slaty structure was developed readily in wax by warming it, kneading it, and pressing it between thick wet plates of glass. If the wax is next cooled, for instance in a freezing mixture of ice and salt, its slaty cleavage can be developed readily with the point of a penknife ("Fragments of Science," D. Appleton, N. Y., 6th edition, vol. 1, p. 316). It is of course the flow and not the pressure that causes the fibrous or slaty cleavage under these various conditions, quite as the flow under the rolling pin causes the slaty cleavage of pastry and biscuits, perhaps by distributing the butter present in thin films which later serve as internal surfaces of weakness, yielding when the enclosed air is expanded on heating the mass.

[2] See Table 4, §50. 1 or even 2 per cent. of slag by weight, or 2 to 4 per cent. by volume, is common in wrought iron, whereas Stead gives the maximum of sulphide areas in steel as not more than 0.5 per cent. (*Fifth Report Alloys Research Comm.*, excerpt *Proc. Inst. Mechan. Engin.*, 1899, p. 77). This was evidently not intended to include such extreme cases as that of Brinell's steel of 0.56 of sulphur and about 1.5 per cent. of sulphide, No. 21, Table 39, Appendix II.

[3] *Journ. Iron and Steel Inst.*, 1887, No. I, p. 262.

[4] "The Closing of Blowholes in Steel Ingots," the Author, *Proc. Amer. Soc. Testing Materials*, 1909, vol. 9, p. 327. Stead gives evidence tending to show that a segregate, rich in phosphorus and sulphur, often squeezes into these blowholes, and this of course would not be removed by the welding (*Proc. Cleveland Inst. of Engineers*, 1912–1913, Fig. 4 after p. 51). Some care is needed to distinguish segregates thus squeezed into blowholes from interdendritic segregates remaining in their initial position.

[5] Stead, *Journ. Iron and Steel Inst.*, 1900, No. II, vol. 58, p. 60; Brearley, "The Use of Microscopic Methods in Relation to Industrial Problems," *Proc. Sheffield Soc. of Engineers and Metallurgists*, 1909–10, pp. 56 to 59; Rosenhain, "Physical Metallurgy," 1914, p. 303; Stead, *Journ. Iron and Steel Inst.*, 1915, No. I, vol. 91, p. 140; *Engineering*, 1915, vol. 99, p. 563. For methods of detecting the phosphorus banding see Table 37, Appendix II.

are usually impoverished in carbon and hence enriched in ferrite—indeed they are often almost pure ferrite—while the dark ones are enriched in carbon and hence in pearlite, or at least in cementite. Moreover, the light bands are usually richer in phosphorus than the dark ones, and this enrichment, which is due to the local concentration of the phosphorus in the interdendritic spaces during solidification as explained in §778, appears to be the usual cause of the banding of the ferrite and pearlite. These ferrite ghosts very often enclose slag masses for reasons which will be given in §780. Some of the ferrite ghosts shown in Fig. A2 of Plate 3 are seen clearly to be made up of distinct ferrite grains of the usual type, like those of Fig. A of that plate.

The usual ghost formation is evidently a variant of the network structure, the assembling of the pro-eutectoid ferrite about the austenite grains in cooling through the transformation range, a ghost being a thickening of the ferrite along certain streaks. A combination of ghosts and network is shown in Fig. H of Plate 44.

The readiness with which appreciable banding is developed varies so greatly as to give the impression that it is only certain steels that are subject to banding. It is probably more accurate to say that the degree to which banding occurs varies so greatly that many steels may be regarded as practically bandless. But much further study of this whole subject is needed.

769. Influence of Phosphorus Content on Ghosts.—Stead[1] gives it as an "indisputable and established law that the micro-ghost lines in steel bars increase in number and extent with rising phosphorus." Yet these ghosts may be so prominent in steel so nearly free from phosphorus as to imply some other cause, such as the banding of manganese, silicon, nickel, or chromium.[2] Thus the prominence of the ghosts of Fig. G of Plate 43 in steel of 0.014 per cent. of phosphorus is out of proportion to that in Fig. C with 30 times as much phosphorus. So with the broad ferrite bands of Fig. A2 of Plate 3, in steel which, by an analysis of drillings taken from this spot, contained only 0.02 per cent. of phosphorus.

It is to be remembered that very slight differences in structure and composition suffice to cause variations in etching tint, so that there may be several types of difference, of which each is capable of causing etching ghosts. The ghosts caused by the banding of phosphorus are those which have been studied most, and they naturally receive our chief attention in what follows.

As will be explained in §775 the banding may, by suitable procedure in casting, be made very narrow, so as to be removed the more easily by diffusion aided by rolling or forging.

770. The Size of the Ferrite Ghosts.—In large forgings which have undergone relatively little drawing out these ferrite masses are often so large that they show very plainly to the naked eye as the forging is machined,

[1] *Journ. Iron and Steel Inst.*, 1915, No. I, vol. 91, p. 149.

[2] See Le Chatelier and Lemoine, *Comptes rendus*, 1915, vol. 161 p. 373.

and of course without any etching. Because they are commonly rod-like in such forgings, any individual band is likely to be removed bodily by the lathe-cut next after that which has made it visible, and to be replaced by another in a neighboring place. Hence their name "ghosts." In smaller sections, such as those of rails[1] and rods,[2] they are usually seen easily by the naked eye on polishing and then etching deeply enough to remove the surface which has been flowed by the polishing itself. The ferrite ghosts in steel of 0.46 per cent. of carbon are shown without any magnification in Fig. D of Plate 43.

771. Ferrite Ghosts in Steel and Slag Banding in Wrought Iron.—The ferrite-pearlite banding in steel is in a way like the ferrite and slag banding in wrought iron, for in each case the weak and ductile ferrite has its continuity broken up by a more brittle substance. Yet the two cases are very unlike in other respects. Though in very low-carbon steel the ratio of size which the ferrite bands bear to those of pearlite may be not far from that which the ferrite bands of wrought iron bear to those of slag, in medium-carbon steel this ratio is very different. Here the ferrite bands may be very much narrower than those of pearlite (Fig. H of Plate 44), whereas in wrought iron the ferrite bands are incomparably wider than those of slag. Moreover, the pearlitic masses of steel, though less ductile than those of ferrite, yet may have very important ductility, and are usually very much stronger than the ferrite bands, whereas the slag of wrought iron is not only perfectly brittle but extremely weak. Finally, the preventability of ghosts distinguishes them sharply from the inevitable slag streaks of wrought iron.

772. The Origin of Fiber. Dendritic Segregation.—The dynamic cause of fiber is the banding of the structure caused by the elongation which the mass undergoes in rolling, forging, etc. But in order that this drawing out should cause banding, there must be something to band, some chemically or structurally different bodies in the metal capable of being drawn out, such as the slag, blowholes, and the phosphorus concentrations already considered. These differences arise during the solidification of the molten steel, which proceeds differentially as described in §§198 and 774 by the deposition of solid layers increasing progressively in richness. The carbon content of the solid metal in the act of deposition or solidification is at first about half that of the initial molten metal, and it increases progressively as solidification proceeds, till it may at last become much greater than the initial average of the whole, causing macroscopic upper-axial segregation or enrichment of the last freezing parts.

This deposition is of a dendritic growth along the shores of the solidifying mass, which comes to resemble a frozen submerged moss, with its minute trunks and branches *gd* and *ef* of Fig. 112 representing the first solidified metal poorest in carbon, and the fillings between, such as the space *ab*, representing metal later to solidify and hence richer in carbon.

[1] Stead, *op. cit.*, p. 26, and *Proc. Journ. Cleveland Inst. of Engineers*, 1905–1906, p. 164.
[2] The Author, *Trans. Amer. Inst. Min. Eng.*, 1913, vol. 47, p. 621.

773. Reasons for the Dendritic Form.[1]—An equiaxed salt crystal, suspended in the middle of a brine which is barely saturated, tends to remain equiaxed by growing at the same rate in all three directions, essentially because the conditions are nearly alike in all three.

But when such a crystal grows inward from the sides of the vessel as in Fig. 112, its tip tends to grow faster than its sides because it adjoins liquor richer than that which bathes the sides. This is because the growth of the crystal impoverishes the liquor out of which it grows, and because the re-enrichment of that liquor by diffusion is more rapid at the tip which protrudes into the open sea than along the sides which face a landlocked harbor. When growth has reached the stage sketched in Fig. 112 the tip has a manifest advantage over the sides as regards the re-enrichment of its environing liquor by diffusion from the great open reservoir in the middle of the vessel. Whatever enrichment of the liquor is to occur at *K*, to feed the little side branches there, must be by diffusion along the narrow inlet *ba*.

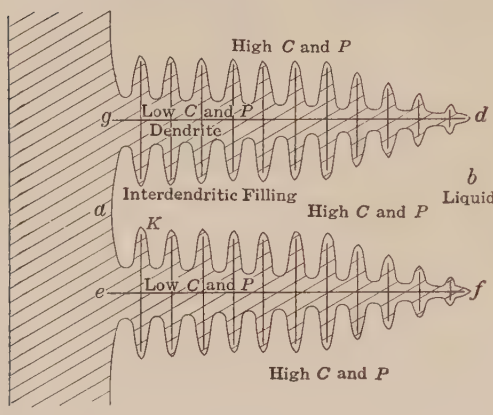

FIG. 112.—Dendritic structure during solidification.

This principle which at this stage leads to the rapid growth of the tips has already led to the formation of the pine trees themselves. Any slightest protrusion, at any one spot on the surface of the solid layers at their contact with the brine, has a corresponding advantage over the concave spaces about it. Because it protrudes it will the sooner be supplied by diffusion with liquor rich enough in salt for it to feed on. Hence from being a slight protrusion it grows rapidly into being an outshooting blade. Because it thus grows faster than the concavities beside it, it continuously increases its advantage. "To him that hath shall be given."

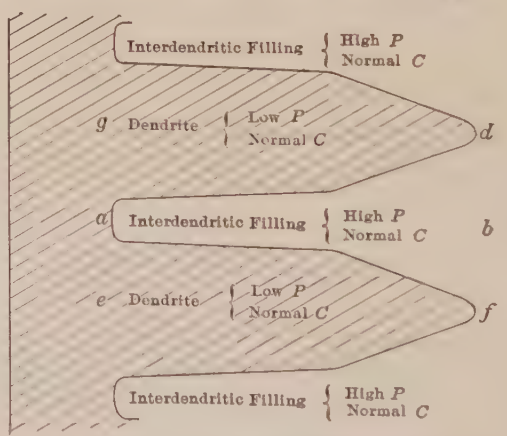

FIG. 113.—Dendritic structure after cooling slowly to a temperature a little above the transformation range.

This spindling habit here replaces the equiaxed habit as a result of the greater stimulation of the tip than of the sides, quite as in the case of trees which are broad and bushy in the open, but in a forest shoot up because their tips are better sunned than their sides.

[1] See §§107 and 343.

774. Solidification of the Landlocking Type Opposes, that of the Onion Type Favors, Upper-axial Segregation.[1]—At the opposite extreme from this dendritic or landlocking type of solidification is the "onion" type, in which the deposition is that of a series of smooth layers, approximately concentric with the mould. In solidification of this latter type the concentration of the carbon, phosphorus, and sulphur which occurs in the deposition of each successive layer would naturally push these segregating elements forward into the last freezing region, which is usually in the upper part of the axis and immediately below the pipe.

But dendritic solidification, by the development of closely packed dendrites, landlocks much of these impurities, and thus distributes them relatively evenly in a vast number of minute pockets between the dendrite trunks. Hence the slightness of axial segregation in quiet and therefore undercooling and dendritic steel.[2]

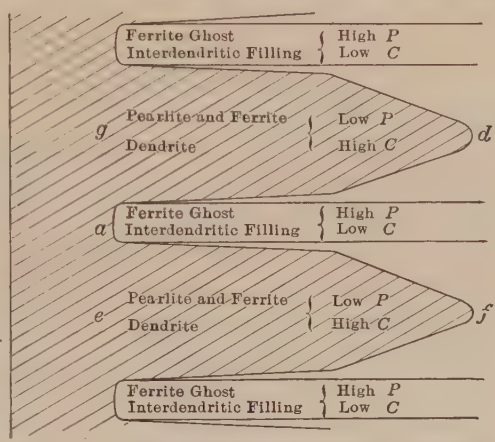

FIG. 114.—The banded structure formed in cooling through the transformation range.

This local or landlocking type of segregation is much the less harmful, because the unevenness of composition represented by each of these little local segregates is far more completely removable by diffusion, aided by the kneading of rolling and forging, than the large-scale unevenness of upper axial segregation. The closer the packing of the dendritic trunks, that is, the closer the individual trunks stand to each other, the smaller, first, will be the content of the carbon and the several impurities in the richest point of each local segregate, and, second, the more readily will the heterogeneousness of composition represented by this segregation be broken up by diffusion and mechanical kneading, both because of the lessened degree of local enrichment and because of the lessened distances to be covered in turning heterogeneousness into homogeneousness. Hence, though we prefer the dendritic form of segregation in order to avoid the harm done by the upper axial segregation to which the onion type of solidification leads, and though the dendritic form is the cause of fiber, we aim to lessen this fiber by pushing the dendritic type to its extreme, making the dendrites as numerous and hence as small as possible.

775. The conditions which favor the dendritic form are rapid and quiet solidification. Rapid solidification increases the advantage which the tips have by giving diffusion little time to bring fresh brine to the branches

[1] The Author, *Iron, Steel and Other Alloys*, pp. 85–99, Sauveur and Whiting, Boston, 1903.

[2] See the Author, *Year Book Amer. Iron and Steel Inst.*, discussion of E. F. Kenney's paper, 1915, to appear. Also "General Principles of the Control of Piping and Segregation in Steel Ingots," *idem* (October, 1915). Also *Iron Age*, 1915, vol. 95, p. 1244, and 1915, vol. 96, p. 995.

near the bases. By increasing the number of nuclei from which solidification proceeds independently, it leads to the close packing of a great number of small dendrites,[1] rather than to the development of a few broadly branching ones.

Quiet favors the dendritic form, both by its lack of convection currents which would bring fresh brine up to the base branches, and also by leading indirectly to rapid solidification by inducing undercooling, that is, the failure of solidification to start till the temperature has fallen well below the liquidus. When the solidification of an undercooled solution once starts, it goes on very rapidly. This effect is seen easily and strikingly in the solidification of a solution of sodium sulphate, saturated when hot, and cooled quietly with the complete exclusion of dust and all other nuclei which might start solidification. Under these conditions it cools far below its true liquidus. When at last solidification is started, for instance by dropping in a grain of the salt, it goes on very rapidly, by the outshooting of these pine trees which I have been describing, or rather of radial blades of this same general form, evidently because of tip-stimulation. For like reasons it is probable that long thin dendrites of frost on the window pane, or of ice on the pool, represent rapid cooling and stillness.

These principles apply equally to the solidification of a metallic mass. Among the conditions which lead here to rapid solidification, and thus to the close packing of the dendrites, to small scale of the dendritic segregation, and to lessening of the upper axial segregation, are low casting-temperature, which avoids storing in the mould walls much heat which is later available for retarding the solidification; massive cold moulds which extract the heat rapidly; casting in narrow ingots, which cool rapidly; early stripping; and above all quiet, through thorough deoxidation, to induce undercooling.

The fact that a high casting temperature leads to a coarse columnar fracture in small ingots both of carbon and of manganese steel, like but less marked than that of Fig. 15, p. 82, though the fracture of like ingots of the same steel, allowed to cool before casting, may not be columnar at all, I refer to these same principles. After the lower casting temperature the dendrites are so closely packed that the fracture passes across rather than between them. The coarse dendritic fracture of the "scalded" ingot results from the fewness of the dendrites and their consequent great individual size, and the weakness of their contacts.

Turning to intelligent practice, these benefits of casting in small ingots are thought to outweigh the accompanying decrease in the kneading or work which, under these conditions, can be put on the metal before it reaches its final shape. One of the most intelligently managed tire establishments, which formerly made its tires from relatively large ingots, has returned to the use of single-tire ingots on this account, finding that the small and

[1] Compare Tammann, "Kristallisieren und Schmelzen," Barth, Leipzig, 1903, p. 131.

quickly solidifying ingot, with its slight dendritic segregation, gives a better tire than the large ingot, in spite of the smaller total work put on the metal.[1]

Again, after an American establishment which has the very highest reputation for the excellence of its products had installed an admirable slabbing mill for rolling large boiler plate ingots, it, too, returned to the use of small ingots (§804). The crucible steel makers, in making the best of all kinds of carbon steel, in like manner cast in small ingots and in what at first seem needlessly thick-walled iron moulds.

776. The columnar structure in the outer parts of steel ingots, especially in those of quiet steel, may be regarded as dendritic, and may according to these principles be referred to rapid solidification. So with the columnar outer parts of chilled cast-iron castings.

Two reasons suggest themselves for the abruptness with which the long continuous columns are replaced by completely equiaxed grains, with no intermediate stage, as shown in Fig. H of Plate 18.

The columns themselves may represent the undercooling stage of the solidification. The quiet metal cools well below its solidus before solidification actually starts. But when this once begins it goes on rapidly till the heat evolved by the solidification brings the shore layer of the molten up to the liquidus, and thus ends the undercooling. This presence of the solid layers prevents undercooling from recurring, so that henceforth the solidification is of a lower order of rapidity. This retardation of the solidification might be so great and so abrupt, and the solidification henceforth so slow, that the advantage which the pine-tree tips have over any other nuclei is insufficient to lead to appreciable inequiaxing of the grains which now solidify.

An alternative explanation is that the end of the columnar region represents the layer at which the progressive concentration of the dissolved gases in the molten reaches effective supersaturation, so that they start to escape, and that the convection currents which they create break up the quiet which has led to the columnar deposition.

Each of these two mechanisms, which indeed do not exclude each other, might bring about the abrupt change which we notice from the columnar to the equiaxed structure.

777. The Dendritic Form caused by the Undercooling of Eutectics and Eutectoids.—This same principle is illustrated by the dendritic form of eutectics and eutectoids, for instance pearlite. Because there are no convection currents; because of the general sluggishness of the mass; because in hypo-eutectoid steel there are no nuclei of cementite present to incite the precipitation of pearlitic cementite at the temperature at which it is due, $Ae1$; because therefore the actual precipitation of cementite is deferred to a much lower temperature, $Ar1$, the lower the faster the cooling; and because therefore the pearlitic cementite, once it starts to precipitate,

[1] "Rolled Steel Wheels," Catalogue No. 12, 1913, of the Standard Steel Works Company, Philadelphia.

forms with very great rapidity, it takes a true dendritic form, that of very broad thin plates. We may question whether pearlite ever has the lamellar form, which represents this assembling of the cementite in sheets, except when the transformation is rapid, because of undercooling or for some other reason.

778. Ghosts Caused by the Interdendritic Concentration of Non-ferrous Matter.—The important aspect of the dendritic solidification for our present purpose is its landlocking, and so concentrating in the spaces between the pine-tree trunks, part of each of the non-ferrous substances present, the carbon, phosphorus, sulphur, slag, and gases most prominently, but probably also to a minor degree the manganese and silicon. This concentration tends toward completeness in the case of slag, because this is practically insoluble in the metal, but as regards these other substances it is progressive, each concentric layer being a little richer in each non-ferrous element than that deposited last before it. But in spite of this progressiveness, for our present purpose each space between the axes of neighboring tree-trunks may be divided up into two parts, the lower carbon parts nearer those axes, hatched and marked "dendrite" in Fig. 112, and the higher-carbon interstitial ones, marked "interdendritic filling." For simplicity we may here ignore the saw toothing of the branches sketched in Fig. 112, and regard as the dendrite the whole space enclosed within its general outline, as hatched in Figs. 113 and 114. In the case of manganese we have yet to learn whether the interdendritic fillings are richer or poorer than the dendrites.

779. In cooling slowly from the solidus toward the transformation range the concentration of the carbon tends to efface itself rapidly through diffusion, but that of the slowly moving phosphorus, and probably of the maganese and silicon, is far more persistent, so that, on arriving at the transformation range after a slow cooling, the state shown in Fig. 113 is reached, the carbon content being now relatively uniform, but great concentration of the phosphorus in the dendrites persisting.

In further cooling through the transformation range, pro-eutectoid ferrite is formed as usual, but much more abundantly in the fillings than in the dendrites, so that though the initial distribution of the phosphorus is retained, that of the carbon is reversed, as sketched in Fig. 114, for this ferrite enrichment of the fillings means the carbon enrichment of the dendrites. It is thus that the ferrite ghosts come to represent the interdendritic fillings, drawn out by rolling or forging into longitudinal threads in rails and like objects, and into thin laminæ in rolled plates and sheets, while in castings they form milky ways. It is because the fillings are richer in phosphorus than the dendrites that they generate a larger proportion of pro-eutectoid ferrite.

780. Slag Enclosures in the Ghosts.—The ferrite ghosts often enclose slag masses, as follows from their stratigraphic history. The concentration of the slag in the fillings during solidification is unlessened by diffusion in any later stage. Hence when the ferrite concentrates into the

fillings in cooling through the transformation range it finds the slag already there, and assembles about it, somewhat as shown in Fig. F of Plate 43 and G of Plate 44 (see §367).

781. Evidence for this Hypothesis of Ghost Formation.—In specimens subject to banding, the persistence of the phosphorus banding is shown by etching with Heyn's[1] or better Stead's or Le Chatelier's[2] cupric reagent, which develops lighter and darker bands, the former richer, the latter poorer in phosphorus, whether the heat treatment has been such as to cause or to efface the ferrite banding, provided only that it has not included so extremely protracted a heating to far above 900° as would at last diffuse even the phosphorus (Table 37, p. 601).

Stead's reagent has been shown to detect the banding not only of phosphorus but also that of nickel, copper, arsenic, chromium, tin, and antimony, brought about like that of phosphorus by the differentiation of solidification. It is probable that it discloses also the banding of still other elements, such as manganese and silicon. In the same way it darkens ferrite more than cementite. The darkening by the cupric reagent by the preferential deposition of copper on the more electro-positive ferrite, leaving the more electro-negative cementite dark, differs radically from the darkening by picric and nitric acids of the pearlite, and in general of the parts less rich in ferrite, by whatever mechanism this occurs. Yet in cupric etching also a darkening effect parallel to this latter may occur, and not only mask but even reverse the expected darkening of the more electro-positive parts.

Turning to picric acid etching, the tint of which is much darker on pearlite and sorbite than on ferrite, but is relatively little affected by the phosphorus content, the diffusion of the solidificational carbon banding during slow cooling toward the transformation range is shown by the absence of banding in specimens quenched after holding above *A*3. The assembling of the ferrite in the fillings in cooling through the transformation range is shown by the progressive increase in the prominence of the ferrite bands, coinciding in position with the phosphoric bands, as the cooling through this range is retarded from experiment to experiment. The etching of course is done in all cases after complete cooling.

Finally, ferrite banding which has been suppressed by rapid cooling through the transformation range is developed gradually by reheating to temperatures below that range, and is then gradually reabsorbed by carry-

[1] Heyn, *Mitteilungen a.d. kgl. Materialprüfungsamt*, 1906, No. 5, p. 253. Heyn and Bauer, "Metallographie," Göschen, Leipsic, 1909, vol. I, p. 21. Martens-Heyn, "Materialenkunde," IIA, Springer, Berlin, 1912, p. 178.

[2] *Proc. Cleveland Institution of Engineers*, Session 1914–15, Nos. 2 and 3, p. 73; and *Journ. Iron and Steel Inst.*, 1915, No. I, vol. 91, p. 173, and also pp. 166–167. (See Table 37). The copper deposits faster and thus has the greater darkening effect on the purer than on the less pure parts, and in general on the more electro-positive substances, leaving the more electro-negative bright. But manganese is reported to have the opposite effect, hastening the precipitation of the copper Le Chatelier and Lemoine, *Comptes rendus*, 1915, vol. 161, p. 378.

ing the heating into and past that range, and then quenching the specimen so as to fix the *status quo*.

782. Lightness Referable in Part to Impurity.—For simplicity I have thus far spoken as if the streaks which are light under picric acid etching owe their lightness solely to their greater ferrite content. More broadly, they owe it to their etching more slowly than the rest, from any cause, such as different purity from that of the darker bands. Hence great caution is needed in interpreting these etching phenomena. Those who lack great experience and skill may easily be misled.

783. The persistence of the phosphorus banding under various conditions has been shown by Stead.[1] Thus whereas he has shown that on a greatly prolonged exposure to very high temperatures, as in cooling down in a large mass of molten slag, the phosphorus banding is effaced, he finds that it is present in steel cooled so rapidly from 1,000° that no carbon banding occurs (his Plate XXXV, Figs. *k* and *l*). A 4-day cooling from 1,200° diffuses the phosphorus so far that a single phosphoric band now covers the space initially occupied by an alternation of several sets of phosphoric bands and of purer bands; yet this diffusion leaves several of the initial purer bands still relatively free from phosphorus, with the result that the number of ferritic and of pearlitic bands has decreased and the width of each pair correspondingly increased (his Plate XXXV, Fig. *c*).[2]

This same persistence is shown indirectly by the redevelopment of ferrite bands after they have been effaced, if we admit, as is extremely probable, that it is this persistence that makes this redevelopment possible in most cases.

784. The spreading of the carbon by diffusion above A3 is shown by the usual absence of banding from the picric acid etchings of specimens which, after a sojourn of even moderate length above *A*3, are then cooled rapidly enough to prevent the differentiation thus effaced from recurring in passing through and below the transformation range (Plate XXXV, Fig. *k*). I have verified this by finding the ghosts effaced completely on heating to 1,100° and quenching. It is true that Stead shows slight banding in his steel of 0.135 per cent. of carbon and 0.061 per cent. of phosphorus after quenching heating to 900° and quenching, and very faint banding even when the temperature was 920° (*idem*, Figs. *i* and *j*), though the *Ac*3 for steel of this carbon content should lie at about 880°, and that I find many ghosts in steel of 0.46 per cent. of carbon and 0.041 of phosphorus when quenched after heating for 10 minutes to 800°, and a few ghosts after like heating and quenching at 1,100°. Yet this persistence of banding may be interpreted with some confidence as implying that there has not been a sufficient sojourn truly above *A*3, either because the stay was too short or because

[1] *Journ. Iron and Steel Inst.*, 1915, No. I, vol. 91, p. 168. The figures referred to by means of small italics in parentheses in this and the immediately succeeding sections refer to this paper.

[2] Stead, *idem*, p. 170.

*A*3 was raised locally by the phosphorus content to above 920°.[1] This latter explanation cannot apply to the last case, the 1,100° quenching.

785. The development of ferrite bands by cooling slowly through the transformation range is shown by the banding which picric or nitric acid causes thereafter. The prominence of this banding increases with the slowness of cooling, as is shown by Figs. A and B of Plate 43. Stead shows such banding in steel of 0.135 per cent. of carbon after slow cooling from 1,000°, 890°, 850°, and 800° (Plate XXXV, *f*, *g*, *h*, and *o*), and Oberhoffer[2] found that this banding reappeared in boiler plates after rolling from 1,200° to 885°, from 900° to 760°, and even to below *Ar*1, and after annealing at 1,000°, 850°, and 750°. There is no reason why it should not.

In industrial practice, repeated and even protracted annealings followed by slow coolings have, to my knowledge, failed to remove transverse weakness, which led to the rejection of gun members. This evidence is less cogent, because we do not know certainly that this specific case of transverse weakness was due to ferrite banding.

The development of these ferrite bands in cooling occupies a very appreciable time, as is shown by their absence from Stead's plate of 0.135 per cent. of carbon when air cooled from 1,000°, and their progressive increase in prominence as the time of cooling was lengthened to 10 minutes, to half an hour, and to an hour (*idem*, Figs. *k*, *m*, *n*, and *o*). More strikingly still, even a cooling which occupied 20 minutes in passing from 900° to 600° failed to develop them in one of Stead's experiments although the phosphorus was strongly banded. (*Idem*, Plate XXXVII and p. 168.)

786. The development of ferrite bands by reheating quenched steel is shown in Figs. E and F of Plate 43, representing two steels which had been freed from ghosts by diffusing the carbon by heating above the transformation range, and then quenching so as to preserve the *status quo*. They were then reheated to 600° and cooled slowly, when they showed these ghosts.

This redevelopment of ghosts on reheating steel in which they have been effaced by quenching from *A*3 is shown more fully by my experiments with steel of 0.46 per cent. of carbon and 0.041 per cent. of phosphorus. A series of specimens were first made ghostless thus, and reheated to an ascending series of temperatures, varying by short steps, and again cooled slowly. No ghosts were seen when the reheating reached 550° or any lower temperature, but they appeared when it reached 600°, and they then in-

[1] Howe and Levy found that the presence of 0.417 per cent. of phosphorus raised *Ac*3 from about 820° to about 870° if the initial ingot structure was removed, and even to above 950° if it was not, or at the rate of about 1.2° and 3° for each 0.01 per cent. of phosphorus. The smaller effect after the removal of the ingot structure is probably due to lessening the phosphorus peaks by diffusion. If the phosphorus were completely equalized by diffusion, its effect in raising *A*3 would probably be much less (*Trans. Amer. Inst. Min. Eng.*, 1913, vol. 47, p. 603). Stead also has noted this raising effect of phosphorus (*Journ. Iron and Steel Inst.*, 1915, No. I, vol. 91, p. 149).

Zeits. anorgan. Chem., 1913, vol. 81, p. 156 and Plates I to IV.

creased markedly in prominence as it was raised by 25° steps to 650°, but only slightly if at all during the further rise to 750°.[1]

Here, as in cooling through the transformation range, the coalescence of the ferrite into visible bands needs time. Thus the relatively faint ferrite ghosts of my steels of about 0.15 per cent. of carbon when air cooled grew progressively in prominence when the air cooling was succeeded by reheating to 700°, holding there for 4 minutes, 9 minutes, and 27 minutes, and then air cooling.

If the reheating of the quenched steel is followed by requenching instead of by slow cooling, then though the banding naturally increases in prominence as the requenching temperature rises say from 550° to 650°, quite as in the cases just considered, yet as the temperature rises into the transformation range and the pro-eutectoid ferrite is therefore correspondingly reabsorbed, the banding in the requenched steel becomes progressively fainter, again as in the cases just considered in §784, evidently because the quenching denies the opportunity which slow cooling gives for the ferrite, reabsorbed in heating into the transformation range, to reprecipitate.

The ghost development recorded in this section and the preceding one represent the coalescence of ferrite in the phosphoric parts at temperatures at once far enough below *A*3 to bring the ferrite into existence and high enough to permit migration.

787. The Mechanism of the Ghost Formation.—We have seen that the banding of the phosphorus is caused by drawing out, in rolling or forging, the dendritic segregates which have been enriched in phosphorus during solidification, and that where the phosphorus is concentrated, there the ferrite concentrates itself both during slow cooling through the transformation range, and on reheating steel which has been quenched and thus denied the opportunity for banding during cooling. By what mechanism does this concentration occur?

In the case of slowly cooled steel, because the ferrite of these ghosts, like the pro-eutectoid ferrite of the network structure of ghostless steel, comes into existence during the cooling through the transformation range, and hence is to be regarded as pro-eutectoid ferrite equivalent to the network ferrite of ghostless steel, but differing from it in being concentrated in longitudinal phosphoric bands, we infer that the ghost ferrite, like that of the network, is expelled by the mother austenite which gives it birth. But though the mother austenite thus does this expulsion, the surface tension of the ferrite itself may play a very important part in determining the shape of the grouping which that ferrite adopts.

The phenomenon which we observe is that, both in the slowly cooled steel, and in that which after rapid cooling is again heated to 600°, there is more ferrite in the phosphoric bands than in the less phosphoric ones,

[1] The prominence differs so greatly in different fields in a single small specimen that no firm opinion of the temperatures at which it reaches its maximum can be given without thorough study. Steel of 0.40 per cent. of carbon and 0.014 of phosphorus gave results much like these.

though there is no corresponding difference in carbon content between these sets of bands in quickly cooled steel. Whence this difference? Is it that the phosphoric mother austenite generates more ferrite than the rest, or that the ferrite migrates[1] from the less phosphoric bands into the more phosphoric ones, and thus raises their ferrite content? Let us consider these two explanations.

788. Does Phosphorus Increase the Generation of Pro-eutectoid Ferrite by the Mother Austenite?—This, the most evident hypothesis, seems incompetent, because an air cooling, though it certainly ought to give time for the completion of the chemical transformation from austenite into ferrite and cementite, and hence to generate in the phosphoric bands the greater ferrite content which this present explanation calls for, yet causes no appreciable ferrite banding. Certainly if the apparently greater quantity of ferrite which the light bands of a furnace-cooled specimen suggest, as in Fig. B of Plate 43, is really present at the end of an air cooling, it ought to be visible then and to cause like banding. But I find that in many cases an air cooling is too brief to cause banding. Because the banding is not visible after it, but becomes visible on prolonging the opportunity for coalescence, either by retarding the cooling or by reheating the air-cooled or quenched specimen, though without increasing the total quantity of ferrite, we naturally seek next some coalescence explanation of the banding.

789. Does Phosphorus Cause Banding by Accelerating the Coalescence of the Ferrite?—This it might do in two ways, indirectly through its known effect of raising the transformation range, and directly by making the ferrite more mobile. Let us consider these two explanations.

790. Effect of the Raising of A3 by Phosphorus.—We have just seen in §784 that phosphorus raises materially the temperature of *A*3, that at which the mother austenite begins generating pro-eutectoid ferrite in slow cooling. Hence this generation of ferrite, in slow cooling, will begin earlier in the more phosphoric bands than in the purer ones.

Beati possidentes!—In coalescence, as in the closely related phenomena of crystallization and grain growth, to him that hath shall be given. The ferrite expelled later by the mother austenite might well, in the course of its expulsion, coalesce with the first formed nuclei.

This coalescence is the result of surface tension between the ferrite masses which, even if unrecognizable in our microsections, constitute nearly the whole of the mass, for instance 94 per cent. and 98.5 per cent. respectively

[1] Stead presents this from the opposite point of view, that of the migration of the carbon away from the phosphoric bands. Within the transformation range the separation is of ferrite from austenite, below *Ar*1 it is of ferrite from pearlite. If I am right in referring the movement to the greater mobility of the phosphoric ferrite, then it is rather the ferrite that leaves the pearlite than the reverse. This conception has the advantage of linking the phenomenon with the genesis of the ferrite network by expulsion from the mother austenite grains, which is described more aptly as a removal of the ferrite from the austenite than of the austenite from the ferrite, as in child-birth it is rather the child that moves away from its mother than the mother from the child (§§367 to 369).

of the mass of steel of 0.40 and 0.10 per cent. of carbon. Here, quite as in grain growth and crystallization, the larger masses always tend to draw to themselves the smaller ones and thus to absorb them. This relative fixity of the large ferrite masses is all the greater because they are anchored by the slag bodies which they so often surround.

The earlier precipitation of ferrite in the phosphoric bands starts ferrite nuclei to which the later precipitated ferrite from the less phosphoric bands may attach itself in its migration, as it is expelled by the mother austenite. Slower and slower cooling permits the more and more complete assembling of the ferrite from the purer bands about the already established larger ferrite masses in the phosphoric bands.

791. The Ferrite Banding caused by Reheating Quenched and Bandless Steel.—The foregoing explanation may suffice to explain the ferrite banding which occurs during slow cooling, but it fails to explain the banding of ferrite on reheating quenched or air-cooled steel, which after such rapid cooling shows no banding of carbon. Let us consider the more cogent of these two cases, that of the banding which occurs on reheating quenched steel. In this case (1) we naturally expect the carbon not to be banded in the quenched steel, and the evidence indicates that it is not; yet (2) it evidently is banded when the steel is reheated; and hence (3) this banding is naturally referred to the migration of the ferrite into the phosphoric bands on reaching the temperature at which the mobility is such as to permit this migration by coalescence. Let us consider these three matters.

792. The Quenched Steel has no Carbon Banding.—The carbon is not banded in the steel when above the transformation range as we have seen in §784. The rapid passage through the transformation range on quenching should not lead to banding of the carbon either through the suggested increase of the eutectoid carbon content by phosphorus or through the known raising of $A3$ by phosphorus, first because it denies the time needed for the migration which such banding implies, and second because, in a 0.40 per cent. carbon steel, it restrains the transformation so far as to prevent the generation of pro-eutectoid ferrite. Here we must remember that in such quenched and reheated steel there is neither pearlite nor pro-eutectoid ferrite strictly speaking. The differentiation which, in passing slowly through the transformation range, generates those two constituents, is prevented here by quenching, so that there is no single particle either visible or conceivable of which we may say that it is pro-eutectoid ferrite or that it is pearlitic ferrite.

Turning from deduction to induction, we have good evidence that the carbon is not banded in the quenched steel. If it were, then when such a steel is reheated so as to enable the transformation into ferrite and cementite to complete itself, as it would even at a temperature as low as 400°, the bands poorer in carbon would contain more ferrite than those richer in carbon and would hence be lighter in shade. But no such difference in shade occurs either on reheating to 400°, or even on reheating to 550°,

though at this latter temperature the transformation has had far more than the needed opportunity to occur.

793. The Banding caused by Reheating Quenched Steel Represents Greater Mobility of the Phosphoric Ferrite.—Because banding, representing a greater proportion of ferrite in the light than in the dark bands, occurs when this steel after reheating to 550°, is further reheated to any temperature between 600° and *Ac*1, and because no known transformation can occur in this range, we naturally refer this banding to coalescence. Because these light bands coincide with the phosphoric bandings, we refer this excess of coalescence in these lighter bands to the greater mobility of the phosphoric ferrite.

On this theory, the more rapid coalescence on reheating such a quenched specimen would assemble both ferrite and cementite into larger masses in the phosphoric than in the purer bands, and hence would cause the phosphoric bands to look lighter, in part because of the greater prominence of their larger ferrite particles, but more particularly because their coarser cementite particles would decompose less and hence darken less under the etching reagent than the finer cementite particles of the purer bands.

794. The Greater Mobility of the Phosphoric Ferrite Contributes to the Banding of the Ferrite in Slow Cooling.—Having found this explanation needed and sufficient to explain the ferrite banding which occurs on reheating quenched steel, we see that this principle, once admitted, must come into play in slow cooling through the transformation range. Hence, though the banding which occurs in this cooling may be due in part to the raising of *A*3 by the presence of phosphorus, it is to be referred in part to the greater mobility of the phosphoric than of the purer ferrite. Indeed, this latter principle seems the more important of the two.

These two explanations thus sufficing to explain the phenomena, the suggested increase of the eutectoid carbon content by phosphorus becomes superfluous. These explanations are offered provisionally pending fuller study of the facts.

795. The Incompatibility Theory. Phosphorus Retards Cementation.—Some seem inclined to explain these phenomena by a supposed incompatibility between carbon and phosphorus, so that austenite migrates away from the phosphoric bands when within the transformation range, and that cementite does when below that range. In support of this the slower cementation of phosphoric than of non-phosphoric iron may be cited.[1]

But this seems to me rather a statement than an explanation, because it does not connect the phenomena with other and familiar ones, and this after all would seem to be the essence of an explanation. One hardly sees why the slower cementation of phosphoric parts is closely related to the ghost phenomena of the greater abundance and prominence of ferrite in those parts. To invoke an incompatibility seems to be proceeding by homology rather than by analogy.

[1] *Journ. Iron and Steel Inst.*, 1915, No. I, vol. 91, p. 171.

Moreover, there is no such incompatibility between carbon and the phosphoric austenite, because during exposure to any temperatures between *A*3 and the solidus the carbon, if previously banded and thus concentrated into the less phosphoric parts, quickly becomes evenly distributed, which implies that it migrates from the less into the more phosphoric parts.

796. Persistence of Dendritic Segregation.—Because there is nothing to cause the spreading out and equalizing of slag bodies and unwelded blowholes, these features of the dendritic segregation may be expected to persist indefinitely, though of course the greater the reduction by rolling, the narrower and more discontinuous they should become. Because the diffusion of phosphorus is extremely slow, the ferrite ghosts may be expected to be extremely persistent in slowly cooled steel even after this has been rolled or forged. Note the dendritic segregation structure shown in the rolled steel wheel of Fig. C of Plate 15, and in particular the curved flow lines brought out by the etching.[1]

The dendritic distribution of the phosphorus thus persists long after the initial austenite has transformed into ferrite and cementite in slow cooling and has formed grains which are equiaxed and hence wholly irrelated to these dendrites. As a result the phosphorus content varies from side to side of any given grain in accordance with the initial dendritic solidification, so that these two structures, the dendritic solidificational structure and the equiaxed structure of, say, the ferrite, are superposed, as shown, for instance in Fig. H of Plate 44.

Further evidence of the persistence of this dendritic structure is given by the difference between the longitudinal and transverse properties discussed in §§811 *et seq.*

797. Blowholes an Alternative Cause of Ferrite Ghosts.—If blowholes form, and if their formation is succeeded by severe pressure while the landlocked interdendritic fillings are still soft enough to flow under this pressure, they may in this way be squeezed into the neighboring blowholes. Like the interdendritic fillings, the blowhole fillings should be enriched initially in the non-ferrous elements, especially in carbon, phosphorus, and sulphur, and during cooling from the solidus to *Ar*3 they should lose their carbon rapidly by diffusion, but their phosphorus and probably the other elements only very slowly. In ingots and other castings these two classes of fillings may be distinguished from each other by their position and shape, the interdendritic fillings naturally being longer than the blowhole ones. But in rolled objects this may be much more difficult.

The width and level of the top of the pipe mark the stage in solidification when the contraction of the inner part of the walls and of the molten,

[1] For extremely instructive examples of the persistence of this ingot structure through rolling and heat treatment, see Engineer N. Belaiew, *Rev. de Metallurgie, Mem.*, 1912, vol. 9, p. 647. A convenient reagent for developing this dendritic structure is 1 part of hydrochloric acid, 3 parts of sulphuric acid, and 9 parts of water, used boiling for from 5 to 60 minutes. The etching, after washing, may be inked with printer's ink and printed on common paper with a common letter press. Fig. C of Plate 15 is photographed from such a print.

taken jointly, begins to exceed that of the outer part of the walls. This width and height are often so great as to indicate that this stage comes very early during solidification. From this time on the inner part of the walls should remain under tension, thus lacking the pressure which we have imagined to fill the blowholes with segregate. But a certain wave of expansion passes slowly across the solid walls of the ingot, starting when the very outer shell passes $Ar1$. This expansion of the subcutaneous layers, after this wave has passed beyond the outer and most rigid ones, together with the pressure from the evolving gas may give the motive force needed for filling the blowholes.

CHAPTER 32

INFLUENCE OF MANUFACTURING CONDITIONS ON FIBER

Summary.—Having considered in Chapter 31 the underlying causes of fiber, and the life history of the ferrite ghosts, let us study here the disposal of fiber and its quantitative influence on the mechanical properties.

Because the most abundant evidence has to do with plates, the direct, slabbing, and universal modes of plate rolling are outlined (§§799 to 805). The inevitable longitudinal disposal of the fiber in rails, bars, and wire is the best (§806). The means of fitting the disposal of fiber to the conditions of service is studied in §806. The last passes in rolling have a preponderating influence on the direction of fiber (§807) and frequent alternations of the direction of rolling should reduce the degree of fiber (§810). The actual differences between longitudinal and transverse properties are studied in §§811 to 823, from the data assembled in Table 30. The transverse deficit of tensile strength is much less than that of elongation, and in the case of plates this in turn is less than that of contraction of area (§§812 to 815). An explanation of these differences is suggested in §816.

The fiber in steel ingots is transverse, instead of being longitudinal as in rolled and forged objects (§817). The inequiaxing of the ferrite grains by cool rolling is thought not to contribute greatly to fiber (§821). The fiber is greater in thin than in thick slabbed plates, but no such difference is found in unslabbed ones (§822). Opinions as to the permissible difference between transverse and longitudinal properties are given in §823.

The influence of sorbitizing and of ductility in increasing fiber is considered in §§825 to 827, the industrial occurrence of fiber in §828, and the influence of the mode of rupture on the fibrousness of the fracture in §§829 and 830. The intentional incorporation of slag and manganese sulphide into molten origin steel is considered and deprecated in §§832 and 833.

798. The Mitigation of Fiber.—In addition to bringing about the close packing of the dendrites by quiet and rapid cooling as explained in §775 so as to lessen the intensity and the scale of the dendritic segregation, we may combat directly the specific forms of this segregation, the banding of slag, of blowhole segregates, and phosphoric ferrite. Among the most important means are mechanical kneading and outspreading by rolling and forging.

The presence of slag may be lessened by care in casting, especially by arranging matters so that the metal shall be quiescent in the casting ladle, and hence by making most of the final additions in the furnace, and by making the slag extremely fluid. To this end the addition of alkaline substances in the casting ladle has been proposed.

The prevention of blowholes is brought about by freeing the metal

thoroughly from dissolved gases and especially from oxygen, by deoxidizing additions such as silicon, aluminum, and titanium, before casting it, and by casting at the lowest practicable temperature. The best deoxidizing agent should be first that which is most thorough, and second that which forms the lightest, the most fluid, and hence the most swiftly coalescing compound with any other slag-making matter present, such as suspended slag and any unreduced oxides of iron and manganese. The coalescence of this slag-making matter into particles of appreciable size leads to its removing itself by rising to the surface. Aluminum and titanium seem to be far more thorough than silicon and manganese.

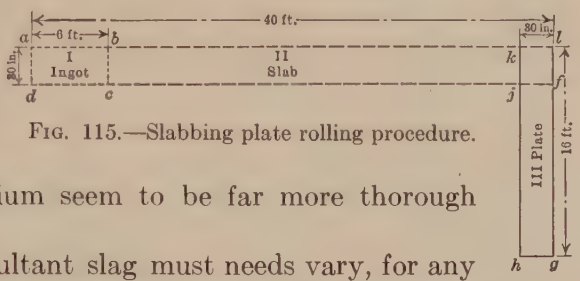

FIG. 115.—Slabbing plate rolling procedure.

The composition of the resultant slag must needs vary, for any one kind of addition, with the quantity of oxide to be removed, which in turn must vary from heat to heat. Direct experiments are needed urgently to show what combination of these additions will, under the probable range of oxygen content of the metal, yield the most thoroughly self-expelling slag. The thoroughness with which blowholes, and their result, segregation, have been suppressed in this or that heat by one or another deoxidizer is weak evidence.

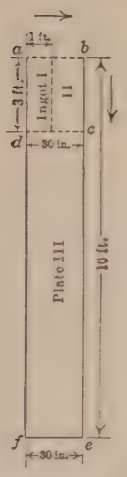

FIG. 116. —Direct plate rolling procedure.

Ferrite ghosts may be lessened by lessening the phosphorus content, and by diffusing their underlying cause, the phosphorus banding, by very long annealing at a very high temperature, of course below the solidus. The coarse granular structure thus induced should then be broken up by reheating from below to slightly above the transformation range.

799. Mitigation in plate rolling may here be considered, in order to illustrate the general principles. The rolling and forging of bars, rails, and most other objects is wholly longitudinal, and hence whatever fiber results from the drawing out of the dendritic segregation is longitudinal solely. But in rolling plates and sheets we can extend the mass both lengthwise and crosswise, and indeed we can arrange matters so as to cause even greater extension crosswise than lengthwise. Let us consider briefly the chief methods of plate rolling in order to get a general conception of the conditions.

NOTE TO FIG. 115.—The ingot *abcd* is first rolled lengthwise into a slab *alfd*, which is then sheared transversely. Each of the pieces of this slab, such as *lfjk*, is then rolled out transversely of the ingot, into a plate *lghk*. Or the slab *lfjk* may be widened out parallel to the length of the ingot before it is lengthened in the way shown in the sketch.

NOTE TO FIG. 116.—The ingot, I, is the first rolled out transversely to the shape II, *abcd*, and then lengthwise into the plate III, *abef*. This procedure admits of great variations, such as cornerwise rolling, and repeatedly changing the direction of rolling from longitudinal to transverse and back, till the length of the plate becomes too long to pass with its length parallel to that of the rolls. Even then it can be rolled cornerwise.

The methods which permit both transverse and longitudinal extension are "direct rolling" and "slabbing." In direct rolling the steel is cast in small ingots, each of which is rolled into a single plate or a group of plates. This leads to "group" that is "bottom" casting if the steel is prepared in large heats, because casting the whole of a large heat into single ungrouped ingots would take so long that the metal would be unduly cold before all were poured, unless indeed it were unduly hot when the first were poured.

In slabbing practice the steel is cast in large ingots which are rolled out lengthwise, and then cut up crosswise into slabs, which are then usually rolled crosswise of the ingot.

The usual procedure in each method is shown in Figs. 115 and 116.

800. In direct rolling the ingot is usually flattened out, say from a width of 12 in. to one of 30 in., or whatever the final width of the plate is to be before trimming. The piece is then turned 90° and rolled to a length of say 15 ft., or whatever is sought, the initial thickness of the ingot being such that when the plate has thus been rolled out to its desired width and length it will have the thickness sought. In fact it has to be rolled to true thickness and then trimmed to length and width. In this practice a single ingot is used for each large plate or group of small plates.

801. In slabbing, a very large ingot, say 6 ft. long and 21 in. by 30 in. in cross-section, is first rolled out lengthwise, as sketched in Fig. 115, and is then sheared crosswise into "slabs," say 30 in. wide, or whatever the final width of the plate is to be. These are then rolled out transversely of the ingot, or partly transversely and partly lengthwise, to the desired thickness, the dimensions of the slab having been such that the length sought will be reached simultaneously with the thickness sought.

In these two cases the procedure is alike in one important respect, that the drawing out is first in one direction and then in another at 90° to the first. Moreover the proportion of lengthwise to crosswise elongation by rolling can be made the same in direct rolling as in slabbing, so that there is no inherent reason, in this respect, why slabbing should lessen the fiber, the difference between longitudinal and transverse properties. For a comparison of the two methods it is easiest to assume that the size of the ingots in direct rolling is the same as that of the slabs in slabbing.

802. Advantages Claimed for Slabbing.—The more intelligible advantages claimed for slabbing are:

1. That the administrative planning is easier.
2. That the slabs can be chipped cold to remove surface defects.
3. That the slab croppings are large enough to be sold to better advantage than the plate croppings.
4. That there is greater total reduction, and hence greater work, put on the metal in rolling.

Slabbing is certainly the more wholesale operation of the two. All the ingots may be of the same size, and all may be rolled out to slabs of the same thickness and size, whereas direct rolling requires that we plan out

the sizes of the many ingots in such a way that each ingot, like each slab in slabbing practice, shall give one plate, or a known number of small plates. Moreover, this planning is made more difficult by the need of having all the ingots of each bottom-cast group of the same length in direct rolling.

The direct rollers may say that this means only that the slabbers avoid the temptation to cast in unsuitably shaped single-plate ingots, and the care which that temptation throws in their way, and that, unless the size of the ingots in slabbing practice is pre-fitted to the weight of the individual plates, large croppings from the slabs must be made. Thus the saving in planning is had only at the cost of large croppings, and hence great discard from the slabs.

803. Advantages Claimed for Direct Rolling.—Against these are to be weighed the advantages claimed for direct rolling:

5. That the casting in smaller ingots leads not only to less axial segregation, but to a smaller scale of dendritic segregation, and thus to better steel.

6. That it saves the cost of reheating the slabs.

7. That it avoids the slabbing discard.

8. That it involves less liability to the external defects called "snakes," cracks which form in the solidifying ingot, and are not readily rolled out in plate rolling because there is no rolling on the edges of the piece, as there is in rolling rails, billets, and other pieces whose width is of the same order of magnitude as their thickness. This liability is less in direct rolling because of the slighter stresses on the skin of the little ingots in solidifying, probably because of their more rapid solidification. By the time that the ingot is so nearly full that the ferro-static pressure on its lower part is severe, the walls have there become so thick that they can resist it. And so with its mid-length.

804. Discussion.—The direct rollers may reply to claim 2 of the slabbers that, because they prevent the formation of snakes they do not need to chip; and to claim 4 that the net segregation, both axial and dendritic, is less with their effective way of preventing it than with the slabbing way of encouraging it by using large ingots and then trying to palliate it by greater total reduction.

This may be summed up by saying that the advantages claimed by the slabbers are administrative, whereas those claimed by the direct rollers concern also the quality of their product. The economies of one practice probably nearly balance those of the other.

Only experts can decide between these rival claims. The claims of the direct rollers seem the more persuasive, especially because they are reinforced by the striking fact that one of the most intelligently conducted plate works, noted for the excellence of its product, though it has a very good slabbing mill, never uses it, but instead rolls direct (§775).

805. Universal plates are those rolled on Daelen's universal mill, the essence of which is that it has, in addition to the usual horizontal rolls,

others with their axes vertical. These latter can put a certain pressure on the edges of the plate, and thus not only prevent the tearing which occurs if this pressure is lacking, but also give the plate an exact predetermined width for its whole length. "Universal" refers to the mills being adapted easily to giving the plate any desired width within wide limits, by varying the distance between the vertical rolls. As an example, ingots 5 ft. long are rolled thus into plates about 100 ft. long, with a reduction in width of between 2 and 4 in.

With this mill the rolling is necessarily almost all lengthwise of the plate, and it is naturally lengthwise of the ingot, lest the imperfections at the ingot-top be strung along one of the edges of the plate instead of being concentrated at one of its ends, and thence readily sheared away. Because the width of plates as they leave the universal mill is that which they are to have in service, whereas that of those rolled on other mills has to be given by shearing, these latter are called "sheared plates."

806. What Disposal of the Fiber is the Best?—What disposal of the elements of fiber, such as ferrite banding and slag masses, will best fit the product for its service? Clearly this must depend on the uses and shape of the product and the stresses which it has to resist. For marine, military, motor car, and aerial purposes, for which every possible lightening of the structure by lessening its sources of weakness is important, the principles which govern the disposal of the fiber may be of moment, for making the direction of greatest strength coincide with that of greatest stress.

The stresses in eye-bars and in most rods are strictly longitudinal, and here the best disposal is exactly that which naturally occurs, drawing out the fiber elements into very long thin rods which offer the minimum of cross-section to the stress.

In the case of plates which have to resist stresses equally lengthwise and crosswise, but not in the third direction, the best disposal is to spread the slag, etc., out equally lengthwise and crosswise.

In the special case of boiler plates, and other cylindrical shapes in which the transverse stress is double the longitudinal,[1] it might be thought well to spread the slag, etc. only circumferentially of the boiler and to avoid spreading it longitudinally, so that it would present the minimum of cross-section to the transverse stress. The practical difficulty arises that the needed width of the plate is usually greater than that which can conveniently be given to the ingot, so that some of the rolling has to be used for widening the plate, and hence only part can be used for the longitudinal drawing out of the fiber elements. But this longitudinal distribution of the fiber is brought about to a very great extent, because the widening is given in the very early passes, and the lengthening in the later ones, and, as we see in §807, the direction of these latter passes is the direction of the chief resultant fiber.

When equal stress in all three directions is to be expected, the ideal

[1] This is familiarly recognized by single-riveting the circumferential seams but double-riveting the longitudinal ones.

disposal of the slag, etc., would be to leave it in masses as nearly equiaxed as possible, so as to restrict its weakening effect toward stress in all directions. This is not easy. A possible way is to roll a large ingot out into a very thick square bloom, cut this into slabs the least dimension or thickness of which is parallel to the axis of the ingot, and then spread this alternately in the two directions normal to the axis of the ingot. The technical difficulties in the way of this equiaxial distribution of the fiber are certainly very great.

An obvious limitation to spreading far in two directions is set by the width of the rolls between housings, which, I believe, does not exceed 12 ft. in any American mill. Hence though the ingot may be drawn out to any desired length, it cannot be spread out to more than 12 ft. in two directions.

The fact that it has been found possible by careful cross rolling to leave the longitudinal strength very nearly equal to the transverse does not in itself imply that the sources of local weakness are negligibly small, but only that they have been balanced and thus concealed.

807. The Last Passes should have a Preponderating Influence on the Direction of Fiber.—Returning to the case of those plates in which we seek to distribute the fiber equally lengthwise and crosswise so as to lessen the difference between lengthwise and crosswise properties, our first thought is that this should be accomplished by making the total crosswise extension in rolling equal the lengthwise. Our first thought is that the dendrites, which run crosswise of the ingot, are widened out by the first longitudinal passes into little plates, which are lengthened and widened by the succeeding longitudinal and transverse passes. That this is true to a greater or less extent is shown by the fact that in both the longitudinal and the transverse section of sheared steel plates the ferrite banding appears as stripes, and not as spots as it does in the cross-section of a rolled rod. Indeed, the longitudinal and transverse sections of these plates are often indistinguishable from each other.

But this conception has to be modified by the tendency of the dendrites, and of the masses into which they are rolled, to break up transversely, as the initial transverse cracks on the surface of the ingot draw out first into longitudinal *V*'s, and then into almost strictly longitudinal streaks.[1] Hence we expect a certain tendency of the little sheets or bands formed during the rolling in any one direction to break up transversely when this direction is changed by 90°, so that the spreading of these little bodies in the penultimate direction of rolling is in part effaced by this cross shattering in rolling in the final direction, and in short so that the direction of the last passes should have a disproportionately great influence in determining the direction of the final fiber. So very competent an observer as Mr. Charles L. Huston[2] informs me that this inference is true.

[1] Compare Wickhorst, *American Railway Engineering Assoc. Report 37 to Rail Commission*, p. 21, June, 1913.

[2] Private communication, Apr. 6, 1915.

808. Evidence that the Influence of the Direction of the Earlier Passes Persists.—However predominant the effect of the final passes in orienting the fiber, they clearly do not efface the direction given in the prior periods. Thus the transverse deficit of both elongation and contraction of area in the tensile test is very much greater in the universal plates, line 26H of Table 30, in which all the rolling was longitudinal, than in *A*, *B*, or *C*, the rolling of which was in three or even four directions, longitudinal, transverse, and one or two diagonal ones.

Again, the transverse deficit of ductility is much greater in plates 26P1 to 26Q4 of Table 30 rolled wholly longitudinally, than in 26R1 to 26R4 finished transversely of the ingot, showing that reversing the direction of rolling has lessened this deficit. Further evidence is given in §822.

The persistence of the influence of the direction of the earlier passes is shown microscopically by the presence of transverse white bandings in Fig. E of Plate 44, the transverse section of the universal plate of the Huston series. This transverse banding must have persisted through the very many longitudinal passes in the final longitudinal direction.

809. The deceptive appearance of flat sections, that is of those parallel to the main surface of the plates, is exemplified by Figs. B and C of Plate 44, in which the direction of the banding, instead of following that of the final passes, is irregular. These indeed are good examples of the pitfalls which beset the interpreter of microscopic appearances. There is a fair correspondence between the transverse and flat sections, Figs. D and E of the universal plate, but none between the patterns on the flat sections *A*, *B*, and *C* and the thin sheet structure of the corresponding transverse sections. Were *A*, *B*, and *C* alone before us we might well infer that the structure in these cases was not lamellar but fagotty, and refer the *V*'s at the left of *B* to a transversed eflection of the previously longitudinal streaks, like the *V*-deflection which the early passes cause in the surface cracks of an ingot, in the process of changing them over to a longitudinal direction.

The explanation is close at hand. A truly plane flat section through a book made up of alternate white and black sheets would disclose one black or one white sheet; but crumple the book or tip the section or make it uneven, and it discloses instead a white and black pattern. Curve either the section or the book, or fold either or both ever so little, and the section will show curved lines as at the right of Fig. C or *V*'s as at the left of Fig. B. It is because of the extreme thinness of the sheets that the slightest departure from planeness of either the book or the section is thus pattern-making. The conditions are even more strongly pattern-making in our sheared plates, *A*, *B*, and *C*. The dark and light sheets of which the plates are composed are seen in Fig. F to be extremely thin and not perfectly regular, and our section made by grinding and polishing may well have deviations from planeness of an order at least as great as that of the thickness of these sheets.

It tallies with this interpretation that throughout the specimens sheared from these plates the patterns are most irregular at the sheared edges, for

first the shearing here distorts the stratification, and second it is at its edge that the specimen becomes most rounded in the polishing.

810. Multiple Reversals of the Direction of Rolling.—By reasoning like that of §807 we see that, whereas prolonged rolling in any one direction should tend to create long bodies of segregate running in that direction, for instance long ferrite bands, and thus to create a marked difference between the longitudinal and transverse properties, or in other words to create marked fiber, frequent changes of the direction of rolling should tend to break these bodies up. This, moreover, should be true of changes from either lengthwise or crosswise rolling to cornerwise rolling.

811. How Great Differences Exist between the Longitudinal and the Transverse Properties?—In order to answer this question I condense in Table 30 the results of more than 1,740 tests or sets of tests combined into 252 groups.

Abundant as this evidence is, it must be used cautiously, because in most cases we do not know either the direction of the greater part of the rolling or for that matter of the final passes, or whether the banding is primarily of the ductile ferrite or of the brittle slag or yet of unwelded blowholes, remembering that the banding of the first of these should differ greatly in its effect from the banding of the others.

812. The Ductility is Affected Much More than the Tensile Strength by Fiber.—Leaving the ingots out of account, because their conditions differ so greatly from those of rolled and forged objects, and confining ourselves to the latter group, we find that in the very great majority of cases, both in wrought iron and in steel, the ductility is affected far more than the tensile strength and elastic limit by the direction of testing. This is exemplified more readily by the objects other than plates, Part II of Table 30, because in plates the rolling is often only in part longitudinal and in important part transverse, so that the effects of the longitudinal drawing out of the dendritic segregation is masked more or less by the transverse drawing out. In these objects of Part II there is indeed a transverse deficit of tensile strength in nearly every case, but, except in two cases, it is very slight, and much less than the transverse deficit of elongation or contraction of area, from one-fourth to one-fortieth as great. It is somewhat so with the plates which we know to have been rolled lengthwise only, Nos. 2 to 5 and 26H of Part I. Here a transverse deficit of tensile strength occurs in only one out of five cases, and even in these it is within the limits of error. Yet in all five the transverse deficit is considerable as regards elongation, and very considerably greater as regards the contraction of area.

The slightness of the deficit of transverse tensile strength is especially striking in the nickel steel, 32 of Tables 30 and 30A. Here the fiber was so strongly marked that a ropy fracture was developed with the greatest ease, yet the transverse deficit of tensile strength is very small.

Turning now to the steel plates of Part I, we find in general but slight transverse deficit of tensile strength. Indeed, though on the average of the

TABLE 30.—COMPARISON OF THE LONGITUDINAL AND TRANSVERSE TENSILE PROPERTIES OF STEEL AND IRON PLATES AND OTHER OBJECTS

No.	Observer	Year	General description	Thickness, inches	Number of groups in which the maximum transverse tensile strength exceeds or equals the minimum longitudinal tensile strength	Total number of groups or cases		No. of groups in which the average transverse properties are less than the longitudinal				Deficit (−) or excess (+) of the transverse over the longitudinal properties, measured in percentages of the longitudinal properties				Loss of elongation divided by loss of tensile strength
						Groups	Cases	Tensile strength	Yield point	Elongation	Reduction of area	Tensile strength, per cent.	Yield point, per cent.	Final elongation, per cent.	Final reduction of area, per cent.	
					I. Molten-origin steel plates											
1	Kirkaldy	1866	Crucible-steel plates	⅜₁₆–⅜	0.187–0.375	4	4	40	1	..	2	1	+ 0.52	.....	+ 7.5	+25.4	....
2	J. Riley	1887	Low-carbon steel	¼–1	0.25–1	.....	56	432	29	..	46	50	+ 0.1	.....	− 6.66	−16.07	....
3	" "	....	The 1 in. thick plates of No. 2	1	1.00	.....	20	...	11	..	15	18	+ 0.1	.....	− 7.9	−12.06	....
4	" "	....	The ½ in. thick plates of No. 2	½	0.50	.....	16	...	8	..	13	14	− 0.45	.....	− 7.77	−17.11	17.3
5	" "	....	The ¼ in. thick plates of No. 2	¼	0.25	.....	20	...	10	..	18	18	0	.....	− 7.82	−16.63	....
6	Gatewood	1886	Low carbon steel	½	0.50	2	2	8	2	..	2	1	− 0.76	.....	−12.48	− 4.27	16.4
7	Barba	1875	" " "	.....	.....	.....	3	28	...	...	...	..	+ 0.2	.....	− 3.61	.....	....
8	A. E. Hunt	1884	" " "	3/16 ±	0.245	.....	1	4	1	..	1	1	− 6.6	.....	− 3.0	− 6.7	0.4
9	" " "	....	" " "	5/8	0.625	.....	1	4	1	..	1	1	− 2.5	.....	− 6.6	− 4.5	2.6
10	Kirkaldy	1881	" " "	¼–1	0.25–1	.....	1	48	..	..	..	..	− 0.3	0	.....	.....	....
11	"	....	The 1 in. plates of No. 9	1	1.00	.....	1	...	..	..	..	..	0	− 0.7	.....	.....	....
12	"	....	The ¾ in. plates of No. 9	¾	0.75	.....	1	...	..	..	..	..	+ 1.4	− 1.3	.....	.....	....
13	"	....	The ½ in. plates of No. 9	½	0.50	.....	1	...	..	..	..	..	− 1.0	− 0.6	.....	.....	....
14	"	....	The ¼ in. plates of No. 9	¼	0.25	.....	1	...	..	..	..	..	+ 1.3	+ 0.5	.....	.....	....
15	H. H. Campbell.	....	Plates rolled from slabs (Penna. Steel Co.)	½–¾	0.50–0.75	.....	1	100	1	1	.	1	− 2.42	−3.27	+ 1.60	− 0.39	....
16	W. Schmidhammer.	1895	Plates from ingot I	.....	.....	.....	1	20	1	..	..	1	− 0.49	.....	− 5.87	− 3.17	12.0
17	" "	....	" " " II	.....	.....	.....	1	20	..	..	..	1	+ 2.76	.....	−10.10	−12.65	....
18	" " "	....	Plates from ingot III from 6 heats	2/5	0.4	.....	1	28	..	..	1	1	+ 1.4	.....	−47.6	−12.0	....
19	" " "	....	Plates from ingot IV	.....	0.48	3	3	30	2	..	3	2	− 0.9	.....	− 7.0	+ 0.3	7.8
20	" " "	....	" " " V	.....	0.48	3	3	30	..	..	3	2	+ 0.4	.....	−24.6	− 3.4	....
21	Kirkaldy	1875	Open-hearth steel plate, unannealed	3/16	0.375	.....	1	2	1	1	..	..	− 0.3	− 0.9	+ 0.4	0	....
22	"	....	" " " " "	1 1/16	0.71	1	1	2	..	..	..	..	+ 0.6	0	+ 0.8	− 2.4	....
23	"	....	" " " " annealed	3/16–1 1/16	0.37–0.70	1	1	12	1	..	1	..	− 0.2	0	− 4.1	+ 3.4	20.5
24	Hill	1880	Open-hearth steel plate	¾	0.75	.....	1	2	..	..	..	..	+ 2.2	+0.3	.....	+12.5	....
25	"	....	" " " "	⅜	0.375	.....	1	2	..	..	..	..	+ 1.6	+0.8	.....	+14.3	....
26	"	....	" " " "	3/16	0.1875	.....	1	2	..	..	..	..	+ 2.1	+0.4	.....	+50.0	....
26A	Wille	1915	Boiler plate	1	1.0	1	1	28	1	..	..	..	− 1.6	.....	+ 3.7	.....	....
26B	"	....	" "	7/8	0.875	1	1	40	1	..	..	..	− 2.5	.....	+ 4.9	.....	....
26C	"	....	" "	1 1/16	0.6875	1	1	60	1	..	1	..	− 2.1	.....	− 6.4	.....	....
26D	"	....	" "	5/8	0.625	1	1	78	1	..	1	..	− 0.9	.....	− 2.9	.....	....
26E	"	....	" "	½	0.5	1	1	104	1	..	1	..	− 1.5	.....	− 6.9	.....	....
26F	"	....	" "	7/16	0.4375	1	1	36	1	..	1	..	− 3.3	.....	− 2.1	.....	....
26G	"	....	" "	3/8 to ¾	0.375–0.75	4	5	30	3	..	1	..	+ 0.9	.....	+ 6.3	.....	....
26H	Huston	1915	Boiler plate, universal mill	1 ¼	1.25	1	1	4	..	..	1	1	+ 2.4	+4.0	−32.0	−51.4	
26I	" (experimental rollings)	....	" " " " "	½	0.5	1	1	8	..	..	1	1	+ 1.8	+0.5	−26.1	−37.0	....
26K	"	....	" " " " "	⅜	0.375	1	1	4	..	..	1	1	+ 1.5	+0.8	−18.2	−24.5	....
26L	"	....	Boiler plate A, Each plate	⅜	0.375	1	1	4	1	..	1	1	− 0.1	+0.1	− 5.3	− 8.8	....
26M	"	....	Boiler plate B, rolled longitudinally, diagonally, and transversely	⅜	0.375	1	1	4	..	..	..	..	+ 1.1	+2.8	+ 2.5	+ 5.8	....
26N	"	....	Boiler plate C	⅜	0.375	1	1	4	..	..	..	..	+ 1.3	+0.4	+ 5.0	+13.8	....

TABLE 30.—COMPARISON OF THE LONGITUDINAL AND TRANSVERSE TENSILE PROPERTIES OF STEEL AND IRON PLATES AND OTHER OBJECTS.—*Cont.*

No.	Observer	Year	General description		Thickness, inches		Number of groups in which the maximum transverse tensile strength exceeds or equals minimum longitudinal tensile strength	Total number of groups or cases		No. of groups in which the average transverse properties are less than the longitudinal				Deficit (−) or excess (+) of the transverse over the longitudinal properties, measured in percentages of the longitudinal properties				Loss of elongation divided by loss of tensile strength
								Groups	Cases	Tensile strength	Yield point	Elongation	Reduction of area	Tensile strength, per cent.	Yield point, per cent.	Final elongation, per cent.	Final reduction of area, per cent.	
26P1	Anon.	About 1915	Plate rolled from ingot exclusively	Tensile strength about 55,000 to 60,000 lb. per sq. in.	⅜	0.375	1	1	9	1	1	1	1	− 0.4	−0.8	− 4.7	− 2.7	11.7
26P2	"	"			½	0.5	1	1	9	1	1	1	1	0	−1.4	−11.6	− 6.6	....
26P3	"	"			¾	0.75	1	1	9	..	..	1	1	+ 3.6	+6.5	−10.0	− 6.6	....
26P4	"	"			1	1.0	1	1	9	..	1	1	1	+ 0.2	−1.9	− 8.7	−14.7	....
26Q1	"	"			⅜	0.375	1	1	9	..	1	1	1	+ 8.4	−0.7	−10.2	− 6.1	....
26Q2	"	"			½	0.5	1	1	9	1	..	..	1	− 0.4	+0.6	+22.6	− 6.5	16.2
26Q3	"	"			¾	0.75	1	1	7	1	1	1	1	0	−0.6	− 7.4	− 8.2	....
26Q4	"	"			1	1.0	1	1	9	..	1	1	1	+ 0.2	−0.6	−27.4	−11.7	....
26R1	"	"	Finished transversely of the ingot.		⅜	0.375	1	1	11	..	1	1	..	+ 0.2	−0.1	−10.3	+ 3.9	....
26R2	"	"			½	0.5	1	1	11	..	1	1	1	0	−0.7	− 3.0	− 8.2	....
26R3	"	"			¾	0.75	1	1	11	..	..	1	1	+ 1.2	+0.1	− 7.0	− 1.6	−
26R4	"	"			1	1.0	1	1	11	1	..	..	1	− 1.0	+0.1	+ 7.8	− 1.8	....
26S	"	1896	American plates		7/16–¾	0.4375–0.75	1	1	68	..	..	1	..	+ 3.6	.....	−11.3	.....	....
26T	"	1897	American basic open-hearth plates		.....	.....	.....	2	...	..	..	1	..	+ 0.1	.....	− 5.5	.....	....
II. Rolled- and forged-steel objects other than plates																		
27	W. Schmidhammer.	1895	Rolled steel bar		⅜	0.4	.....	1	2	1	..	1	..	− 1.0	.....	− 4.1	+ 1.8	4.1
28	" "	....	Annealed steel bar		⅜	0.4	.....	1	2	1	..	1	1	− 0.9	.....	−36.7	−36.7	40.8
29	J. O. Arnold	1908	Mild steel shaft		.....	.....	.....	...	...	..	..	..	..	− 2.9	.....	−15.2	−13.6	5.2
30	Job and Hersey.	1911	Steel tires, failed in service		.....	.....	.....	1	2	1	..	1	..	−12.9	.....	−75.3	.....	5.8
31	" " "	....	" " "		.....	.....	.....	1	2	1	..	1	..	−21.1	.....	−86.4	.....	4.1
32	The Author	1910	Nickel steel, (4.61 Ni, 0.29 C)		⅜	0.375	1	1	6	1	1	1	1	− 2.5	−4.6	−17.6	−15.6	7.0
III. Steel ingots																		
32A	Maitland	1887	Gun steel unforged, heated once	unhardened	.....		.....	1	2	1	1	1	1	− 3.25	−2.85	−30.79	−32.17	10.0
32B	"	....		oil-hardened			1	1	2	..	..	..	..	+ 1.13	.....	+ 8.54	.....	....
32C	"	....	Gun steel, round ingot unforged	unhardened			1	1	2	..	..	..	..	+10.55	.....	+68.11	.....	....
32D	"	....		oil-hardened	.....		.....	1	2	1	..	..	..	− 2.88	.....	+24.33	.....	....
33	Test of metals	1906	Harmet nickel steel ingot	Upper half and mid length	unannealed		2	2	21	..	..	..	..	+27.7	.....	+376.5	+102.9	....
34	" " "	....			annealed		1	1	10	..	..	..	..	+18.9	.....	+42.3	+ 4.6	....
35	" " "	....			forged		1	1	4	..	1	1	..	+ 1.2	− 1.1	− 1.8	+10.2	....
36	" " "	....		Lower half	unannealed		3	3	47	1	..	2	1	+15.3	.....	+ 7.9	+ 1.2	....
37	" " "	....			annealed		3	3	18	..	..	..	..	+26.5	.....	+37.2	+15.8	....
38	" " "	....	Carbon steel ingot	natural state			1	1	6	1	1	1	1	− 0.5	− 3.1	− 6.5	− 9.9	....
39	" " "	....		quenched in various ways			1	1	10	..	..	..	1	+ 6.9	+24.6	+ 8.2	−15.1	....
40	" " "	....	Nickel steel ingot	natural state			1	1	6	..	1	..	..	+ 8.0	− 2.3	+31.7	+53.7	....
41	" " "	....		quenched in various ways			1	1	9	..	..	1	1	+ 3.0	+ 3.0	−14.0	−25.6	....

TABLE 30.—COMPARISON OF THE LONGITUDINAL AND TRANSVERSE TENSILE PROPERTIES OF STEEL AND IRON PLATES AND OTHER OBJECTS.—*Concluded*

No.	Observer	Year	General description	Thickness, inches		Number of groups in which the maximum transverse tensile strength exceeds or equals minimum longitudinal tensile strength	Total number of groups or cases		No. of groups in which the average transverse properties are less than the longitudinal				Deficit (−) or excess (+) of the transverse over the lengthwise properties, measured in percentages of the longitudinal properties				Loss of elongation divided by loss of tensile strength
							Groups	Cases	Tensile strength	Yield point	Elongation	Reduction of area	Tensile strength, per cent.	Yield point, per cent.	Final elongation, per cent.	Final reduction of area, per cent.	
			IV. Wrought-iron and puddled-steel plates and pipe														
42	Kirkaldy	1866	Wrought-iron plates	3/16– 1 5/16	0.19– 0.94	21	26	325	25	..	25	25	− 8.92	.....	−44.45	−42.01	5.0
43	"		Krupp w. i. plates, solid, unannealed			3	3	18	3	2	3	3	− 4.3	−2.1	−31.5	−37.3	8.7
44	"		Krupp w. i. plates, solid, annealed			3	3	18	3	3	3	3	− 2.2	−1.8	−30.1	−38.2	17.8
45	"		Yorkshire w. i. plates, solid, unannealed	7/16, 17/32, 2 1/32	0.44, 0.533, 0.653	3	3	36	3	1	3	3	− 4.7	−0.1	−32.9	−28.6	6.1
46	"		" " " plates, solid, annealed			3	3	36	3	1	3	3	− 4.3	+0.1	−30.4	−23.9	5.6
47	"		Krupp w. i. plates, drilled holes, unannealed	7/16, 17/32, 2 1/32	0.44, 0.53, 0.66	1	1	4	1	..	..	..	− 7.0	.....	.....	.....	....
48	"		" " " punched holes, unannealed			1	1	4	1	..	..	..	− 8.2	.....	.....	.....	....
49	"		Yorkshire w. i. drilled holes, unannealed	3/8– 5/8	0.37 0.39 0.50 0.51	1	1	9	1	..	..	..	− 8.5	.....	.....	.....	....
50	"		" " " punched holes, unannealed		0.52 0.61 0.63	1	1	9	1	..	..	..	− 5.3	.....	.....	.....	....
51	Bauschinger	1882	W. i. boiler plates	.....	.....	.....	1	16	1	..	..	..	−26.0	.....	.....	.....	....
52	"	....	W. i. exploded boiler	.....	.....	.....	1	16	1	..	..	..	−29.0	.....	.....	.....	....
53		....	W. i. six odd plates	.....	.....	.....	1	12	1	..	..	..	−20.0	.....	.....	.....	....
54	Kirkaldy	1866	Puddled steel plate	1/8– 5/16	0.125– 0.312	3	6	37	5	..	6	5	−14.91	.....	−35.68	−38.14	2.5
55	Anonymous	....	Wrought-iron pipe	.....	.....	.....	1	3	1	1	1	1	−28.0	−19.2	− 7.2	− 7.3	0.26
56	"	.....	Charcoal iron plates	.....	.....	.....	1	6	1	..	1	1	− 6.7	+ 5.5	−62.7	−25.5	9.3
			V. Cast iron														
57	A. W. Whitney	1897	Chilled wheel iron I	.....	.....	.....	1	...	1	..	..	..	−41.1	.....	.....	.....	....
58	" " "	....	" " " II	.....	.....	.....	1	...	1	..	..	..	−46.6	.....	.....	.....	....
59	" " "	....	" " " I and II mixed.	.....	.....	.....	1	...	1	..	..	..	−43.7	.....	.....	.....	....

"Solid" here means not punched or drilled.

References to Table 30.—1. Kirkaldy, "Expts. on Wrought Iron and Steel," 1866, p. 145, Table H. Of the 40 tests 22 ran lengthwise, 18 crosswise. 2 to 5. J. Riley, *Journ. Iron and Steel Inst.*, 1887, No. I, p. 144, sheet III, abstract I. All from 12 ingots from a single charge of steel, probably open-hearth. 6. Assistant Naval Constructor R. Gatewood, U. S. N., *Rept. U. S. Naval Advisory Bd. on Mild Steel*, 1886, p. 124, Table XXIV. The pieces were all cut from the same plate, apparently of open-hearth steel of 0.16 per cent. carbon, made at the Chester Rolling Mills. 7. "Use of Steel," Barba, Holley, pp. 26, 29, Tables V, VII. The elongation is given in only two cases, each of which appears to be the average of an unknown number of tests. Twenty-eight results for tensile strength are given, of which two at least represent more than one test. The remaining 26 results are obtained from punched plates. 8 to 9. A. E. Hunt, *Trans. Am. Inst. Min. Engrs.*, XII, p. 315, 1884. The tests were all made on steel from a single open-hearth charge containing 0.15 per cent. of carbon. Of the 8 tests, 4 were made parallel with and 4 across the direction of rolling. 10 to 14. Kirkaldy, Gatewood, *Rept. U. S. Naval Advisory Bd. on Mild Steel*, 1886, p. 133, from Parliamentary Paper C, 2897, London, 1881. 15. H. H. Campbell, "The Manufacture of Iron and Steel," 1907, p. 316. 16 to 20. W. Schmidhammer, *Stahl u. Eisen*, 1895, vol. 15, p. 372. 21 to 23, Kirkaldy, Burr, "Materials of Engineering," 4th edition, 1894, p. 312. 24 to 26. Hill, Burr, "Materials of Engineering," 4th edition, 1894, p. 311. 26A to 26G. H. V. Wille, Baldwin Locomotive Works, Philadelphia, Pa., private communication. 26H to 26L. C. L. Huston, Vice President, Lukens Iron and Steel Co., Coatesville, Pa., private communication. 26P1 to 26T. 26P1 to 26S represent the product of one of the best known American steel mills. 26T is composed partly of the basic open-hearth product of this mill and the acid open-hearth product of another, the acid and the basic steel in this case having very nearly the same ratio of longitudinal to transverse properties. 27 to 28. Same as Nos. 16 to 20. 29. J. O. Arnold, *Engineering*, 1908, vol. 85, p. 600. 30 to 31. Job and Hersey, *Proc. Amer. Soc. Test. Mat.*, 1911, 11, p. 463. 32. I. H. Woolson, *Report*, Jan. 6, 1910. See Table 30A for details. 32A to D. Maitland. "Treatment of Gun Steel", *Proc. Inst. Civil Engin.*, 1887, vol. 89. The Author, "The Metallurgy of Steel", 1st ed., 1890, p. 195. 33 to 41. "Tests of Metals, Etc.," Watertown Arsenal, Mass., 1906, p. 313. 42. Kirkaldy, "Expts. on Wrought Iron and Steel," p. 150. Of the 325 tests, 163 ran lengthwise, 162 crosswise. 43 to 50. D. Kirkaldy, "Experimental Enquiry into the Properties of Essen and Yorkshire Wrought-iron Plates." 51 to 53. Johnson, "The Materials of Construction," 1897, pp. 482–484. 54. Kirkaldy, "Expts. on Wrought Iron and Steel," p. 145. Of the 37 tests, 20 were made lengthwise, 17 crosswise. 55 and 56. From an American pipe mill. 57 to 59. A. W. Whitney, *The Iron Age*, 1897, vol. 59, No. 3, p. 12.

whole there is a slight transverse deficit, there is a slight transverse excess in no less than 27 lines, against a transverse deficit in only 21. In lines 26P1 to 26Q4, with the rolling wholly longitudinal, there is an appreciable transverse excess not only on an average but in four out of the eight lines. Yet in 17 out of the 26 lines in which as regards tensile strength there is a transverse excess, there is a transverse deficit as regards elongation, and in some of these the deficit is very serious, 24.6 per cent. in one case, 47.6 in a second, 32 in a third and 27.4 in a fourth, the third and fourth representing modern practice.

TABLE 30A.—FIBER, THOUGH STRONG ENOUGH TO CAUSE A ROPY FRACTURE, MAY YET CAUSE NO IMPORTANT TRANSVERSE DEFICIT OF TENSILE STRENGTH. (STEEL NO. 66 OF TABLE 39)

No.	Tensile strength, lb. per sq. in.		Yield point, lb. per sq. in.		Elongation, per cent. in 0.35 in.		Reduction of area, per cent.		Deficit (−) or excess (+) of the transverse over the longitudinal properties, measured in per centages of the longitudinal properties			
	Longitudinal	Transverse	Longitudinal	Transverse	Longitudinal	Transverse	Longitudinal	Transverse	Tensile strength	Yield point	Elongation	Reduction
32	106,000	105,680	67,510	44,165(?)	37.2	28.6	64.36	47.32	−0.3	−34.6(?)	−23.1	−26.5
32	108,200	105,380	88,650	94,010	40.0	28.6	66.57	47.95	−2.6	+ 6.0	−28.5	−28.0
32	104,460	100,000	82,170	89,170	34.3	22.9	53.19	87.58	−4.3	+ 8.5	−33.2	−29.2

813. Wrought Iron vs. Steel.—The wrought iron and puddled steel plates of Part IV, which we may group together as of "plastic origin," differ from the molten origin steel plates of Part I in always having some transverse deficit of tensile strength. Though this deficit rises to 26 per cent. in one group of 16 cases and to 29 in another, yet in many cases it is moderate, and in four, Nos. 43 to 46, it is less than 5 per cent. The plastic origin plates differ from those of molten origin also in lacking the important excess of transverse deficit of contraction of area over transverse deficit of elongation which the latter class relatively often has.

But these plastic origin plates are like the molten origin steel ones in that their transverse deficit is far greater in elongation than in tensile strength, more than five times as great in the six out of the eight cases in which a comparison can be made.

It is interesting that the group of plates in which the transverse deficit of tensile strength reaches 29 per cent. represents a boiler which exploded.

814. Why is the Transverse Deficit Greater in Ductility and Especially in Contraction of Area than in Tensile Strength?—I venture to offer a possible explanation which, even though it should not prove valid, may yet lead on to the true one.

In Fig. 117 let *A* represent a specimen taken crosswise of the direction of rolling, and *B* one taken lengthwise, with the ferrite-pearlite banding running across the former but lengthwise of the latter. The elastic limit will be reached in each when the stress on the ferrite masses, which are the weaker, causes the first measurable degree of slip, which should occur earlier in *A*,

with its smaller minimum ferrite areas, than in *B*. The tensile strength, or maximum stress reached per unit of initial section, is reached when the strengthening of the ferrite by the deformation which occurs in the test itself (§§507, 508, 723, 727) ceases to outweigh the simultaneous loss of sectional area. We can understand that the opportunity for this deformation might be less in *A* than in *B*, though were the opposite true, that too would not astonish us. We shall see in §816 a special reason why this banding tends to make the transverse tensile strength greater than the longitudinal.

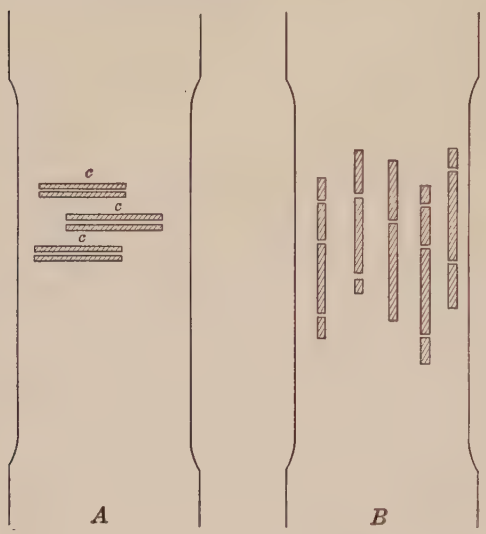

Fig. 117.—The healing inflow of ferrite is easier when the banding is lengthwise than when it is transverse.

But when we come to the ductility the difference in conditions is very marked. The ductility represents the power of the ferrite masses to flow healingly into the cracks which the deformation causes in the inextensible pearlite or slag. With the pearlite bands running lengthwise, as in *B*, and therefore cracking across as sketched there, the inflow of the ferrite to close these cracks is extremely easy and the ductility correspondingly great. But when the bands lie across the specimen, as in *A*, so that they tend to crack open parallel to their length as sketched, the inflow of the ferrite is far more difficult, because the distance to which it has to flow in order to reach the middle of the crack, *c*, is so great. Much the same is true if we suppose the pearlite masses to be rods instead of bands, assuming that the transverse rods will split parallel to their length, and indeed even if we do not suppose them to split at all, but only to oppose, by their stiffness, the transverse flow of the ferrite as it accommodates itself to the progressive contraction of area resulting from elongating the test piece.

815. Elongation vs. Contraction of Area.—Mr. C. L. Huston informs me that the effect of fiber is greater on the contraction of area than upon the elongation, and this is true not only of the six groups in Table 30 representing his practice, but also, with a single exception, of the 28 individual tests from which these four are derived. It is true also of most of the most trustworthy data in Part I of this table, in particular of the excellent Riley plates 2 to 5, in which, as in Mr. Huston's, the banding is very probably of ferrite and not to any important extent of either slag or blowholes. It is not true of the very important group representing modern practice, lines 26P1 to 26R4, or of the objects in Parts II, III, and IV. We may here ignore the cases in which the transverse deficits are so small as to be referable rather to the normal variations from like case to case than to the direction of fiber.

A provisional explanation of this may be derived from Fig. 117. The contraction of area represents the centripetal necking of the specimen after the load passes its maximum. Transverse slabs of pearlite, as in *A* of Fig. 117, act as struts to prevent this centripetal movement, and thereby lessen the tendency to neck, and the contraction of area, much more effectively than like masses do when they lie lengthwise as at *B*.

But the effect on the elongation of this difference in the direction of the pearlite masses need not be so great. The moment that necking starts in, the local cross-section, in the absence of banding, loses so rapidly in area that it is able to transmit to the rest of the test piece only a rapidly decreasing load, so little load that the elongation now ceases in the remainder of the test piece apart from the necking area. But these transverse struts of pearlite in *A*, by opposing this necking, lessen this decrease of the load on the remainder of this test piece, with the result that elongation continues throughout that remainder. Thus the loss of elongation implied by the less contraction of area at the necking section in specimen *A* is offset in some measure by this increase of the elongation elsewhere, with the difference that in specimen *A* the elongation is probably more spread out than in specimen *B*. The strut effect of the transverse pearlite bands, though subtractive of elongation in that it lessens the local necking, is additive in that it increases the elongation elsewhere.

816. The Strut Effect of the Pearlite Decreases the Deficit of Transverse Strength.—These same considerations may help to explain why the transverse deficit is so much less as regards tensile strength than as regards elongation. This strut action of the transverse pearlite, though as we have just seen it is partly subtractive as regards the elongation, is wholly additive as regards the tensile strength, in that it opposes the tendency to neck, that is, the tendency to lessen the existing sectional area and thus to lessen the load which the test piece as a whole can endure, in spite of its progressive strengthening per unit of existing section through continuing deformation. So this strut effect increases the maximum load reached before necking sets in, that is, the tensile strength.

This does not imply that the transverse tensile strength is greater in a banded than in a like but unbanded specimen, but it helps to explain why, in a banded specimen, the transverse tensile strength may be greater than the longitudinal. In other words, this strut effect of the pearlite tends to offset the weakening action of the banding more as regards transverse tests than as regards longitudinal tests, and it tends to offset also the greater weakening effect which would be expected to follow from the presentation of the sides of the banding to the main stress in transverse testing, instead of the ends of the banding as in longitudinal testing. The banding, in that it implies the existence of transverse heterogeneousness in *A*, tends to lessen the strength, to weaken it; in that it implies this strut effect it tends to increase it. The latter effect tends to lessen the former.

There are so many exceptions in Parts II and IV of Table 30 that we

cannot claim that these tentative explanations cover the whole ground. We may indeed refer the exceptions in Part IV to the ineffectiveness of the brittle slag of wrought iron as a strut, but even then we should fail to explain why the deficit of transverse tensile strength is usually so slight.

817. Fiber in Steel Ingots.—Because the dendritic segregation here lies transversely of the ingot instead of longitudinally as in rolled and forged objects, there is not a deficit but a surplus of transverse properties in most cases. This effect is reinforced by the transverse position of the columnar crystals, the junctions between which are surfaces of weakness, as we know from the familiar columnar fracture of "cobbles" from ingots and blooms broken early in their rolling, as shown in Fig. 118. Thus as regards elongation there is either a transverse surplus, often a large one, or a very small transverse deficit in 13 out of 14 groups, representing 125 out of 131 cases. The structure varies so greatly from part to part of these ingots, especially if they contain blowholes, that many exceptions to any rule must be expected.

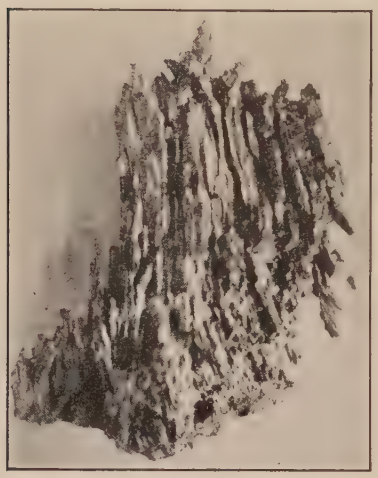

FIG. 118.—Columnar crystals in steel ingot.

818. The Influence of Slight Forging on the Fiber in Steel Ingots.—The effects of such forging, recorded in Table 31, tend to support my inference that the excess of the transverse tensile properties of the ingots over their longitudinal ones is due jointly to the intercolumnar weakness and to the distribution of the dendritic segregation, including blowholes and slag, transversely of the ingot. These sources of weakness present their ends to transverse stress, their sides to stress longitudinally of the ingot. This should make the transverse test pieces cut from the unforged metal stronger than the longitudinal ones, and so we find them in lines 13 to 15 of Table 31. Flattening the ingot cheesewise should exaggerate this excess of the radial over the axial properties, and so it does in lines 16 to 18. But drawing the ingot out lengthwise should draw these sources of weakness out lengthwise of the ingot, so that they will present their ends to longitudinal and their sides to transverse stress, and the longitudinal test pieces should be somewhat stronger than the transverse, and so they are in lines 19 to 21 of Table 31.

819. Chilled cast iron, Nos. 57 to 59, Table 30, is in like manner much weaker across than along the columnar crystals, indicating that here too the intercolumnar boundaries are sources of weakness. In all three cases the strength crosswise of the columns is more than 40 per cent. less than along them.

820. The Harm done by Ferrite Ghosts.—The foregoing discussion has to take the three main elements of fiber, ghosts, slag, and unwelded blow-

TABLE 31.—THE INFLUENCE OF FORGING ON THE RELATION BETWEEN THE LONGITUDINAL AND TRANSVERSE PROPERTIES OF STEEL INGOTS, CALCULATED FROM MAITLAND'S DATA

	Description		Tensile strength, lb. per sq. in.		Yield point, lb. per sq. in.		Elongation, per cent. in 2 in.		Work of rupture, inch-tons per cubic inch	
		Direction of test pieces	Unhardened	Oil hardened	Unhardened	Oil hardened	Unhardened	Oil hardened	Unhardened	Oil hardened
1	Unforged heated once	Longitudinal	72,800	99,126	35,392	.....	21.5	7.25	5.75	.....
2		Transverse	70,504	100,240	34,384	.....	14.875	7.875	3.90	.....
3		Per cent. deficit of No. 2	−3.25	+1.13	−2.85	.....	−30.79	+8.54	−32.17	.....
4	Heated twice, forged once	Longitudinal	71,008	99,904	32,256	.....	27.0	16.75	6.97	.....
5		Transverse	69,328	90,888	31,752	54,432	20.25	10.875	5.12	.....
6		Per cent. deficit of No. 5	−2.36	−9.02	−1.56	.....	−25.0	−35.07	−26.54	.....
7	Heated four times, forged thrice	Longitudinal	74,816	92,512	53,536	59,248	30.0	24.75	9.07	9.00
8		Transverse	70.448	93,072	43,456	58,464	25.5	19.75	6.12	5.35
9		Per cent. deficit of No. 8	−5.84	+0.65	−18.83	−1.33	−15.0	−20.20	−32.53	−40.56
13	Round ingot, unforged	Axially	62,608	77,728	.....	.....	9.25	9.25	.....	.....
14		Radially	69,216	75,488	.....	.....	15.55	11.5	.....	.....
15		Per cent deficit of axial	+10.55	−2.88	.....	.....	−68.11	+24.33	.....	.....
16	Round ingot, flattened	Axially	61,152	91,392	30,464	.....	8.25	6.75	1.87	.....
17		Radially	73,024	97,216	33,376	56,000	26.0	18.75	6.94	7.38
18		Per cent. deficit of axial	+19.41	+6.37	+8.42	.....	+215.1	+177.8	+271.1	.....
19	Round ingot, lengthened	Axially	74,368	87,696	39,200	56,448	29.0	24.25	8.10	8.50
20		Radially	66,080	74,256	30,688	.....	17.75	11.0	5.50	6.90
21		Per cent. deficit of axial	−11.15	−15.33	−21.70	.....	−38.79	−54.64	−32.10	−18.82

1 to 9. Two pieces were cut from the same ingot "so as to be of equal quality." One was forged successively from a section 10 in. × 10 in., to 7 in. × 7 in., to 5 in. × 5 in., and to 5 in. × 2.5 in., each reduction occurring at one heating. The other piece was heated together with the first, but not forged. 1 to 3 give the properties of the second piece heated once but not forged; 4 to 6, those of the first piece heated twice and forged once, from 10 in. × 10 in. to 7 in. × 7 in.; 7 to 9, those of the first piece heated four times and forged thrice, the total reduction being from 10 in. × 10 in. to 5 in. × 2.5 in.

13 to 21. From the upper part of a circular ingot longitudinal and transverse test pieces were cut, Nos. 13 to 15. A part of the same ingot was then flattened down into a cheese, and test pieces were taken transversely and axially, 16 to 18. A third piece was drawn out parallel to the length of the ingot, 19 to 21.

Maitland, "The Treatment of Gun-steel," excerpt *Proc. Inst. Civ. Eng.*, vol. 89, 1887.

holes collectively, because we rarely know which of these is the chief culprit. Yet certain effects of the ferrite ghost structure, the concentration of the ferrite into certain bands and of the pearlite into the alternate ones, can be identified. The ferrite bands when broad are harmful in "dragging" like copper under the cutting tool and thus causing surface roughness. The dark or pearlitic bands are for like reasons harmful in causing brittleness, and when broad in causing such great local hardness that the cuts taken in machining must be lighter than they would otherwise have to be.

Rosenhain[1] finds that the spontaneous cracking of large plates is associated with unusually great banding and variation in phosphorus content between bands. Indeed such heterogeneousness of structure with its cor-

[1] Discussion of Stead's paper, *Journ. Iron and Steel Inst.*, 1915, No. I, vol. 91, p. 186.

responding heterogeneousness of physical properties is harmful in many other ways, for with it neither the weakness of the ferrite nor the brittleness of the pearlite is mitigated as it is in unbanded steel by the intimate admixture of these two constituents.

The ghosts which represent blowhole segregates may at times owe their hardness to an abnormal concentration of cementite and phosphorus. For instance, Edwards[1] describes a case in which the initial phosphorus content of 0.14 per cent. was raised locally to 0.60, and as he infers, to no less than 10 per cent. in the segregate itself, with the result that the steel was too hard locally to be machined. It may well be that those ghosts which are extremely hard represent blowhole fillings as a rule.

Though broad ghosts, involving great heterogeneousness, may do very great harm, narrow and faint ones, such as represent the slight heterogeneousness of well-made boiler plate steel, are so abundant and so usual in excellent commercial steel that the harm which they here do cannot be important.

Like other forms of fiber, ghosts are naturally expected to be most harmful when they lie transversely to the most severe stresses to which the object in which they occur is exposed. But we may suspect that the specific effects of ghosts, like the collective effects of fiber, as inferred from Table 30, though worse as regards ductility when the ghosts are transverse toward the main stresses, yet are not materially worse then as regards tensile strength and elastic limit under quiescent stress than when the ghosts are parallel to those stresses.

Special conditions under which the ferrite ghosts may be harmful are those involving a heating intended to be to just above $A3$, for hardening or for carburizing. Because phosphorus raises $A3$, if for any purpose we heat a ghosted steel only to slightly above what would be $A3$ in the absence of phosphorus, that is, to only slightly above the normal $A3$, we in effect heat not to above but only into the transformation range, with the result that the metal, instead of being wholly austenite, remains ferrite in these phosphoric streaks. The same is likely to be true of ferrite bandings from any cause. Even if $A3$ is not raised, the mere local concentration of ferrite must prolong the time needed for its reabsorption by the austenite. If the steel is now quenched, only the parts which are austenite at the moment of quenching will harden, the ferritic bands remaining unhardened and detracting correspondingly from the effective hardness of the mass, whether for cutting or other purposes.

Or if the steel is carburized at this temperature, while the austenitic bands will absorb carbon rapidly, the ferritic ones will not, with a result recalling that explained in the preceding paragraph, that the steel, if now hardened, will contain parts which are soft, because uncarburized.[2]

Harmful as ghosts doubtless are, the degree to which the damage done

[1] *Ibid.*, pp. 189–190.

[2] See Rosenhain, *Metall. and Chem. Engineering*, 1915, vol. 13, p. 584.

by fiber is due to them as distinguished from slag and blowholes remains to be determined.[1]

821. Does the Inequiaxing of the Ferrite Grains by Cool Rolling Contribute to the Deficit of Transverse Properties?—Three facts go to show that this transverse deficit is not due in any important degree to the inequiaxing or drawing out of the individual ferrite grains, as distinguished from the formation of ferrite banding, parallel to the direction of rolling, by rolling at so low a temperature during the last passes that the elongated grains cannot thereafter become equiaxed.

1. The deficit in question is as great in annealed as in unannealed steel, though annealing removes the effects of cold working nearly or quite completely, including the inequiaxing of the grains.

2. If the deficit were due to this persistent inequiaxing, then it should decrease as the thickness of the plate increases, because with this increase goes a rise of the finishing temperature. But no such relation between this deficit and the thickness of the plates is suggested either by the Riley plates, lines 2 to 5 of Table 30, or by the modern plates of lines 26P1 to 26Q4, all of which were rolled wholly longitudinally of the ingot. We shall see in the next section a special reason why this does not apply to the Wille plates.

3. Such inequiaxing seems to increase the tensile strength as much lengthwise as crosswise. Thus I found the hardness of a tensile test bar of low-carbon steel, the grains of which had been greatly elongated by drawing out in the tensile test, the same lengthwise as crosswise of the specimen.[2]

822. Influence of the Thickness of Steel Plates.—Arranging the tests of 376 plates from H. V. Wille according to the thickness, in lines 26A to 26G of Table 30, we find that, though the transverse elongation has the usual slight deficit in the great majority of the plates as thin as $1\frac{1}{16}$ in. (0.6875 in.), it is slightly in excess of the longitudinal elongation in the plates thicker than this. These plates are believed to have been rolled from slabs. Referring back to Fig. 115, the explanation suggests itself that the cross rolling of the slab II into the plate III transversely of the ingot has not, in these thicker

[1] Stead finds that violent shock transverse to the banding does not start rupture in the ferrite bands themselves if they do not contain a "material amount" of slag or sulphide. Hence he inclines to believe that such bands are not "dangerous or liable to lead to the failure of engineering structures." This inference should be taken cautiously. One aspect of the banding is the formation of brittle pearlitic bands, and in these rupture may well start as it does so often in other pearlitic masses. Here the shock resistance of the piece as a whole may well be lessened by this creation of these brittle pearlitic bands, a necessary consequence of the formation of the white ferrite bands (*Journ. Iron and Steel Inst.*, 1915, No. I, vol. 91, p. 167).

[2] "Are the Effects of Simple Overstrain Monotropic?" *Proc. Amer. Soc. Testing Materials*, 1914, vol. 14, p. 28. The tensile test increased the Brinell hardness of this steel of 0.11 per cent. of carbon from 132 at the grip-ends to 199 in the region of the fracture, but in any given transverse section the hardness was the same lengthwise and crosswise. See also Charbonnier and Galy-Aché, "Etude sur les proprietes physiques et mecaniques des Metaux," *Congres Internat. des Methods d'Essai*, Paris Exposition, 1900, vol. I, p. 270; also Galy-Aché, *Revue de Metallurgie, Mem.*, 1913, vol. 10, p. 587.

This tallies with Thurston's data tending to show that cold rolling strengthens the axial parts of the rolled bar as much as the outer ones (R. H. Thurston, "Reports on Cold-rolled Iron," 1878, and the first edition of this present work, 1890, p. 216).

plates, overcome the influence of the lengthwise rolling from the ingot I to the slab II, with the result that the fiber in them is crosswise of the plate. So, too, there is a change, though an irregular one, from a marked transverse deficit to a considerable transverse excess of elongation as we pass from the thinner to the thicker plates of lines 26R1 to 26R4, in which a corresponding change of direction of rolling is known to have occurred. Here is evidence of the persistence of the influence of the penultimate direction of rolling.

823. What Difference between Longitudinal and Transverse Properties does Good Practice Permit?—In 1884 A. E. Hunt[1] reported a usual transverse deficit of 10 per cent. in tensile strength and 20 or even 25 per cent. in ductility. This represents the rather crude state of the open-hearth steel manufacture at that time. In 1896 Campbell[2] intimated that this deficit could be removed completely by means of slabbing, and gave the data condensed in line 15 of Table 30, but without indicating whether they are picked or representative.

In 1915, C. L. Huston[3] informs me that, though he finds no material transverse deficit of tensile strength in either universal or sheared plates, or of elongation in sheared ones, he yet finds a transverse deficit of 5 to 20 per cent. in the elongation of universal plates, and of about 10 per cent. in the contraction of area of sheared ones.

H. V. Wille[4] believes that the transverse deficit of good boiler plates should not exceed 4,000 lb. per square inch, that is about 7 per cent., in tensile strength, or 3 per cent. in elongation. The competence of these observers is of the highest.

824. The Existence of Fiber in Steel Shown by the Bending Test.—In testing thick plates of Bessemer and open-hearth steel, Webster finds that the tendency of the internal partings to start rupture is less, and hence that the angle bent before breaking is greater, if the generatrix of the bend is parallel than if it is normal to the surface of the plate. Indeed there is even the suggestion that, by facilitating longitudinal shear in the former case, the weak planes lessen the tendency to break in bending.[5]

825. When the Structure is Fibrous, the Tendency to Yield a Fibrous Fracture should be Increased by Sorbitizing the Metal and thus Bringing it to Its Best Structural State in Other Respects.—We have already seen that a brittle wrought iron, which yields a crystalline fracture, is likely to have its slag as much drawn out into fibers before the deformation as a ductile one which yields a fibrous fracture, the fibrous fracture here representing the combination of ductility with fibrous structure.

Like considerations may be expected to apply to steel also. Given the longitudinal distribution of the products of dendritic segregation, phosphorus

[1] H. H. Campbell, "The Manufacture and Properties of Iron and Steel," 1st edition, 1896, The Scientific Publishing Co., N. Y., p. 222.

[2] *Idem*, pp. 223, 349, *et seq.*

[3] Private communications, March and April, 1915.

[4] Of the Baldwin Locomotive Company, private communication, Apr. 26, 1915.

[5] William R. Webster, private communication, Feb. 10, 1914.

banding, slag, and unwelded blowholes, if I am right the metal will have the greatest tendency to yield a fibrous fracture, or in other words it will have the most effective fiber, when it is in its best condition, the most finely divided or sorbitic, or in general when its trans-crystalline strength is so great that the trans-crystalline path, in spite of being the shorter, yet offers greater resistance to rupture than the longer fibrous path along the slag and metal partings. Hence a tendency to yield a fibrous fracture may be offered as evidence that the material is in its best state, or even as evidence of general merit, though from another point of view this is evidence of the presence of longitudinal defects. One of the most intelligently conducted steel companies advertises, perhaps in accordance with these ideas, that certain of its special steels, after case hardening and quenching, are "entirely fibrous"[1] either throughout or in the interior.

826. Evidence of the Fibring Effect of Sorbitizing.—Turning from deduction to induction, we find this inference supported by Brinell's experiments Nos. 26, 34, 42, and 49, Fig. 110.[2] Here the treatment was what I call "grain refining and annealing," and is that which naturally results in a very sorbitic structure. And this is exactly the treatment which resulted in a fibrous fracture.

I have found that this fibrous fracture can be developed both in common carbon steel and in nickel and chrome-nickel steel very readily by this sorbitizing treatment.

A fibrous fracture can be developed also in silicon steel with about 1.5 to 2 per cent. of silicon, by nicking it and bending it slowly, after it has been sorbitized by quenching in thin oil from about 871°C. (1,600°F.) and then drawing back at about 455° to 482°C. (850° to 900°F.), the treatment which the 2-in. by ¼-in. bars used for automobile springs receive.[3]

Again, in the Krupp[4] process of fibring armor plate, this same sorbitizing treatment occurs in the rear of the plate, in which an extraordinarily fibrous structure is developed, so that the plate, when broken by transverse pressure applied to its impact face, and hence with strong longitudinal shearing stress along the rear face, breaks with a fracture like hickory. Were it not for the hardness and metallic color one might take the mass for wood rather than steel.

[1] Soc. Anon. John Cockerill "Electro-S.H.C.," pp. 9 and 10, 1910.

[2] Brinell's fracture *H*, "ash-gray fibrous" forms in all his experiments in which his steel of 0.52 per cent. of carbon, after hardening by quenching from above the transformation range, is next heated to *Ar*1, or about 710°, and then cooled either very rapidly by quenching, or moderately rapidly, treatments which would make its structure sorbitic.

[3] G. L. Norris and G. W. Sargent, private communications, Aug. 12 and Aug. 17, 1915.

[4] United States Patent, Emil Ehrensberger, No. 651,965, June 19, 1900; Emil Ehrensberger, No. 653,413, July 10, 1900; and Albert Schmitz, No. 642,926, February 6, 1900.

The Schmitz patent (p. 1, l. 45) refers to the "highly fibrous" grain of the rear of the plate, which by the process of these patents is integrally united with an extremely hard impact face. It is possible that the fibrous weakness, caused we may conjecture by the presence of iron-nickel-chrome oxide, may in this special case be directly beneficial by enabling the rear of the plate to accommodate itself to the inevitable slight buckling under the impact of the projectile, by sliding in longitudinal shear.

This evidence might at first be thought weak, because this very sorbitizing procedure also leads to ferrite banding, which in itself, as a form of fiber, may well lead to longitudinal shear and a fibrous fracture. Yet this objection loses force when we remember that, though slow cooling from well above the transformation range leads to even more prominent ferrite banding, it does not lead to a fibrous fracture.

827. Ductility a Contributary Cause of Fiber.—Somewhat as the strength of the sorbitic structure tends to make existing fiber more effective, so may ductility, though the exact mechanism may not be traced with like confidence. Certain it is that the most ductile steels yield a fibrous fracture readily.[1]

828. Industrial Occurrence of Fiber.—The Krupp fibrous-backed armor plates are made of chrome-nickel steel, which is precisely that most subject to internal flaws and broominess. These are the natural defects which must be guarded against in this otherwise admirable material, unless, indeed, a certain degree of slatiness is thought beneficial, as in the case of armor plate.

In other industrial cases it is especially in applying sorbitizing treatment to this steel that fiber, in its unwelcome aspect of defective transverse strength, is met. This comes about naturally. In order to meet very trying specifications the excellent nickel or chrome-nickel steel is used, and is brought to its best structural condition by sorbitizing it. This indeed enables it to pass the longitudinal reception tests; but unless the longitudinal slatiness has been prevented, the transverse weakness and brittleness which it causes, not being remedied by the sorbitizing, prevent compliance with the transverse reception tests. Hence the difficulty in meeting the transverse tests for gun-tubes and like objects.

The fact that this transverse weakness and these slaty partings are both characteristics of this steel certainly tends to show that the partings are a chief cause of the transverse weakness. Yet ferrite banding may contribute, and perhaps to an important degree.

In like manner the spreading out of the internal sources of weakness in rolling thin plates for circular saws has resulted in a harmful broominess or slatiness of their teeth.

829. Influence of the Mode of Rupture on the Fibrousness of the Fracture.—Certain tough wrought irons yield a fibrous fracture under favorable conditions, for instance when nicked on one side and bent slowly away from the nick, but a crystalline one under others, for instance when nicked all around and broken with a sharp blow.[2,3] Again, good fibrous wrought-iron armor-plates struck by shot shatter like glass, and with a crystalline fracture.[2] The usual explanation is that during slow rupture the individual

[1] Compare Desch, discussion of Pilkington's paper on "Wrought Iron," *Journ. West of Scotland Iron and Steel Institute*, 1911–1912, vol. 19, p. 302, and Pilkington, *idem*, p. 306, to the effect that it is difficult if not impossible to distinguish certain extremely ductile steel, such as that of the American Rolling Mill Company, from excellent wrought iron by means of its fracture.

[2] Percy, *Iron and Steel*, Murray, London, 1864, pp. 10, 11. I have verified this. Thurston vouches for this effect on armor-plate. *Matls. of Engineering*, Wiley, N. Y., 1898, vol. II, p. 593.

[3] Cf. Bayles, *Trans. Am. Soc. Mech. Eng.*, 1886, vol. 7, p. 270.

grains are drawn out into fibers, while in sudden rupture there is not time for this elongation, and accordingly rupture strikes across the piece, between the crystal faces: perhaps rather a restatement than an explanation (§727).

Thick pieces of soft steel which fail in the bending test usually show a crystalline fracture, though tensile rupture produces in them a silky one.[1] Again, not only do guns, whether cast iron, wrought iron, or steel, whether of brittle or ductile material, on bursting always show a short granular fracture, but Maitland has found that this same fracture always arises when steel tubes are burst by pressure from within, whether this pressure is applied suddenly or gradually, and whether the metal elongates much or little. On the other hand, rods torn in two tensilely by the explosion of gun-powder or even of gun-cotton always yield a silky fibrous fracture.[2] In explanation it is pointed out that, under tensile test of a rod or common test piece, the rupturing stress is in a single direction, and tends to elongate the metal's crystals; that when a tube is burst these crystals are exposed to forces acting simultaneously at right angles, a longitudinal and a tangential stress;[3] the crystals cannot so readily elongate in two directions at once; the tangential stress opposes the tendency of each crystal to elongate lengthwise of the tube, and *vice versa:* hence, although the tube as a whole may elongate greatly, its individual crystals elongate but little. Do they then slide past each other?

Further, a punched soft steel bar yields a crystalline fracture: ream but a knife-blade thickness from the sides of the punch-hole and it yields a silky fracture, rupture in one case apparently starting at the hole's edge and ripping thence—as a ton-strong canvass-roll once notched is ripped by a boy—while in the other case all parts of the section pull jointly. The change, in the regions apart from the hole, is probably due to changed approach of stress rather than changed condition before stress.

830. The Reason for this Influence.—When a bar nicked on one side, so that it may break instead of simply bending double, is bent away from the nick by a succession of light blows, time is afforded, after each fraction of trans-crystalline slip, for the strengthening action to occur along the planes on which that slip has started, that strengthening action which, as we have already seen in §§508, 514, sets in as soon as plastic deformation starts. This strengthening of these first slipping surfaces, relatively to the longitudinal slag and metal contacts, tends to prevent additional slip along them, and thus to divert rupture from its trans-crystalline path to these longitudinal contacts, and thus to cause a fibrous fracture. Hence longitudinal shear along these partings, like the sliding of the leaves of a ream of paper past each other, readily occurs, with occasional stepping across from layer to layer

[1] J. Riley, paper 2236, "The Treatment of Gun-steel," *Proc. Inst. Civ. Eng.*, vol. 89, p. 187, 1887.

[2] Maitland, *idem*, pp. 120–121.

[3] Barlow, *idem*, p. 203. I would point out that in the bending test we have these same conditions, tangential and radial stress acting simultaneously. A fragment of the Pittsburgh 6-in. steel cast gun which I have shows this short granular fracture.

at points where the slag layers are specially discontinuous and the trans-crystalline cleavages specially weak, or specially unfavorably disposed to the transverse shearing stress. Thus comes the fibrous fracture of wrought iron, and its "barking" shown in Fig. C of Plate 40, reminding us of the breaking of fibrous wood.

But when instead of a succession of light blows, a single very strong shock is given, this causes an increase of stress so rapid that there is no considerable time for strengthening along the trans-crystalline planes of first slip, with the consequence that slip continues and propagates along those planes, with only occasional steppings across to parallel ones, and a trans-crystalline fracture results, instead of a fibrous one along the slag and metal contacts, all the more because nicking the bar all round tends to localize rupture in this section and to prevent it from shearing lengthwise of the bar along the slag rods. (§§523, 739).

So the tendency to a fibrous fracture is increased greatly both in wrought iron and steel by nicking on one side only and bending slowly away from the nick. By this means a fibrous steel such as is shown in Fig. I of Plate 42, can readily be made to "bark" in breaking. The outer layers at the convex side separate along the slag and metal partings from the remainder, which bends much further before breaking, much as the bark and the outer layers of a notched bough separate when it is bent.

The ease with which we can make the fracture of such steel fibrous or crystalline at will should make us cautious in interpreting fractures, especially those made in our absence.

831. Illustration of the Influence of the Slag on the Path of Rupture.— That rupture should tend to follow the slag and metal contacts when its general course lies parallel to them is natural enough. But even when its general course is transverse to them it often steps sharply aside in order to follow them locally, as in Fig. B of Plate 41. Here the general path of rupture is vertical, but it steps sharply to the left at AB so as to conform locally with the stratification of the irregular gray slag masses. The way in which these habitually preserve their continuity during rolling even at temperatures as low as 900°, at which the slag itself might have been thought very brittle, is certainly striking.

The slightness of the resistance which these slag masses offer to transverse rupture, and the way in which the ferrite, when under strong pressure, heals up gashes which the rupture of the slag bodies cuts, are made clear by Fig. D of Plate 26. This shows the structure of a cold-punched wrought-iron nut.[1] The long nearly horizontal dark gray islands are slag. Note how the great pressure has forced the white ferrite into even the minute cracks between the angular fragments of slag. The right-hand crevice, at B, though only about 0.000,04 in. wide, or of the order of magnitude of a wave of light, seems to be filled completely with ferrite.

[1] This nut, which is 1½ in. thick, has a hole 1⅜ in. in diameter at top and 1⅛ in. at bottom punched through it, by Messrs. Hoopes and Townsend of Philadelphia, to whom my thanks are due.

832. Intentional Additions of Slag.—From time to time it is proposed to improve molten-origin steel by intentional additions of slag. Let us look broadly at the matter.

Though strength and ductility increase the tendency of wrought iron to break with a fibrous fracture, there is no conclusive reason to expect fiber, whether caused by the addition of slag or by cold deformation, to increase either strength or ductility. Indeed the grains of cold-worked and hence brittle iron are elongated, while those of tough hot-worked iron are equiaxed. Moreover, it is not clear that the fact that the former are not equiaxed has any important direct effect on the properties of the metal, for the strength of cold-worked iron seems as great across as along the fiber.

Again because toughness and slag both cause fiber, some befogged ones infer that slag causes toughness. Health, rouge, and intemperance redden the cheeks: do rouge and rum give health?

These fallacies pricked, let us examine the reasons to expect that slag should toughen iron, and the evidence that it does.

Slag may affect iron both chemically and mechanically. Chemically, the slag of wrought iron may toughen the metal by oxidizing carbon and silicon, for the basic iron silicates of which it consists are energetic carriers of oxygen. In molten-origin steel this action is less important, because the carbon and silicon are better removed otherwise, and because, at least in acid steel, the acid slag has little oxidizing power. Indeed the oxidation of silicon makes matters worse, by substituting silica for it.

But no relation between the percentage of slag and that of carbon in wrought iron can be traced in the results of the United States Test Board, as rearranged in Table 32, lines 6 and 7.

TABLE 32.—ANALYSIS OF THE INFLUENCE OF SLAG ON THE PROPERTIES OF WROUGHT IRON, AND ON ITS PERCENTAGE OF CARBON

Twelve wrought irons numbered according to their tensile strength, etc. (1 highest), and placed in the order of the percentage of slag (highest first)

Number for												
Tensile strength	7	9	13	14	16	5	6	3	15	10	2	1
Reduction of area	14	3	2	16	1	10	6	9	4	5	9	11
Elongation	10	1	4	15	3	8	2	9	6	11	13	16
Welding value	9	4	2	....	3	12	10	6	5	....	13	....
Shock-resisting power	13	7	1	14	4	11	9	6	4	6	12	14
Per cent. of carbon	4	3	11	7	6	5	8	9	10	8	2	1
Per cent. of slag	1	2	3	4	5	6	7	8	9	10	11	12

This table is derived from data in the "Report of the U. S. Board to Test Iron, Steel, etc.," 1881, I, p. 224, Table III.

Mechanically, slag (*a*) breaks up continuity, (*b*) brings the metal a step toward the condition of a wire rope or the leaves of a book, (*c*) slightly hinders rupture from striking straight across the piece, and (*d*) probably retards grain growth.

The first action weakens and embrittles. The second promotes flexibility, but we may question whether it promotes ductility as measured by elongation and contraction of area.[1] On the other hand it should lessen the transverse strength, as the evidence in Table 30 indicates that it does. Note that there is a deficit of transverse tensile strength in each of the lines representing wrought-iron plates of Part IV, but only in 22 lines representing only 39 groups of molten origin steel plates, out of a total of 53 lines representing 175 groups.

The third might be important were the strength of slag comparable with that of iron. Buried steel rods tend to prevent rupture from striking across a concrete mass. But we cannot confidently expect the brittle feeble rods of slag to obstruct the path of rupture greatly, though they naturally give flexibility.

That slag masses do not obstruct rupture greatly is indicated by Baker's experiment in Table 28, in which an artificial crack weakens wrought iron as much as steel. The cracking of the slag masses shown in Fig. D of Plate 26 tallies with this belief. Rupture may indeed be restrained from striking across the piece, but probably less by the resistance of the slag itself than by the lack of continuity due to the presence of slag.

The effect of slag in retarding grain growth in iron remains to be studied. Thoria is incorporated in the tungsten filaments of incandescent lamps for the express purpose of retarding the grain growth which limits the life of the lamp. Here thoria forms little globules, and bears to the tungsten a relation much like that of slag to iron. Nor is the evidence at hand more favorable than these reasons to the plan of adding slag to molten origin steel. The ductility of pure Swedish wrought iron is indeed excellent, but there is no strong reason to believe that this represents its slag content rather than its purity. The ductility of very low-carbon steel, *e.g.*, of the American Rolling Mill Company, almost pure ferrite, may be about as great as that of the best wrought iron, as is indicated by the cases which I collect in Table 33.

The abandonment of the much discussed attempt at Avesta in Sweden to incorporate slag in molten steel is reported to have led to no diminution of the great ductility which had been claimed for this artificially enscoriated metal.[2]

[1] Barba's law that test pieces of similar proportions undergo similar deformations implies that the contraction of area is independent of the diameter, but that the elongation in a given measured length increases with the diameter (compare Unwin, "The Testing of Materials of Construction," Longmans, London, 1910, p. 91). H. H. Campbell's data show that, between the diameters of the $1\frac{5}{16}$ in. and $\frac{5}{8}$ in. the elongation increases but the contraction of area decreases as the diameter increases, in the case of round bars of basic open-hearth steel of carbon 0.11 to 0.15 per cent. and phosphorus 0.02 to 0.04 per cent. "The Manufacture and Properties of Iron and Steel," Hill, N. Y., 1907, p. 323. Here the conditions are affected by the lower finishing temperature, the greater reduction in rolling, and consequently the smaller scale of the dendritic segregation in the thinner bars, and this explains why the contraction of area in Campbell's cases does not obey Barba's law. Moreover, the proportions of his test pieces are not similar, because no change in the length is made to correspond to the change in the diameter.

[2] Fischer, *Oest. Zeitschrift*, 1886, vol. 34, p. 244, 1886. Goedicke, *Idem*, p. 536. Drown, *Proc. Soc. Arts, Mass. Inst. Technology*, 1885–86, p. 150. Raymond, Howe, *Engin. and Mining Journ.*, 1886, vol. 42, pp. 181, 219.

TABLE 33.—UNUSUAL DUCTILITY IN STEEL AND WROUGHT IRON

No.	Material	Tensile strength, lb. per sq. in.	Yield point, lb. per sq. in.	Elongation			Reduction of area, per cent.	Authority	Carbon, per cent.
				Observed, per cent.	In in.	Observed or calculated, per cent. in 8 in.			
1	A.R.M. steel.....	55,720	38,380	24.13	8.0	24.1[b]	79.67	Cushman	Less than 0.02
2	" "	48,900	33,460	44.0	2.0	27.0[b]	68 8	W. J. Beck, A.R.M. Co.	
3	" "	49,300	33,930	49.0	2.0	26.2[b]	61.4	" "	
4	Electrolytic iron, forged.	55,400	46,100	39.5	2.0	23.2[a]	69.5	Burgess, C. F.	More than 0.01
5	Basic mild steel...	56,000	.....	33.0	8.0	33.0[b]	64.7	Pilkington	0.10
6	Wrought iron.....	48,764	32,233	47.0	2.0	27.6[a]	76.5	Arnold	0.04
7	Wrought iron,.. Lowmoor.	48,160	.....	28.5	8.0	28.5[b]	56.9	Pilkington	0.12
8	" " "	51,300	.....	32.0	8.0	32.0[b]	60.0	Richards, E. W.	
9	" " charcoal.	57,216	.....	30.2	6.4	27.0[a]	60.9	Kirkaldy	
10	" " Swedish.	45,014	24,360	22.0	60.0	29.0	77.0	Styffe	
11	" " "	47,855	.....	27.8	6.0	25.2[a]	60.5	Kirkaldy	
12	" " "	48,232	.....	26.4	7.0	25.4[a]	68.0	"	
13	" " pipe, American.	34,890	34,090	36.0	2.0	25.4[a]	44.6	Speller	
14	Charcoal iron plates, American.	46,730	29,540	30.0	8.0	30.0[b]	48.7	"	

Notes to Table 33.—[a] Calculated from data given by Unwin, "Testing of Materials of Construction," 3d edition, London, 1910, p. 91

[b] Observed.

1. *Proc. Amer. Soc. Testing Materials*, 1911, vol. 11, p. 406.

2 and 3. Private communication, July 31, 1915.

4. Private communication, Sept. 5, 1911.

5 and 7. *Journ. West. of Scotland Iron and Steel Inst.*, 1912, vol. 19, p. 259.

6. *Journ. Iron and Steel Inst.*, 1894, No. 1, p. 118.

8. *The Engineer*, Oct. 16, 1896, p. 386.

9. "Experiments on Wrought Iron and Steel," London, 1863, pp. 110–111. Bradley Co. charcoal iron.

10. "Iron and Steel," translated by Sandberg, London, 1869, pp. 132–133. Rolled iron made in a Lancashire hearth from Lesjöforss, Wermland, Sweden.

11 and 12. "Experiments on Wrought Iron and Steel," London, 1866, pp. 116–117. Swedish iron from the Eckman Co., Gothenburg, Sweden.

13 and 14. Private communication, August, 1915.

833. Manganese sulphide, which may here be regarded as the equivalent of slag, is sometimes generated within steel for the purpose of causing it to yield short brittle chips in machining, by breaking up its continuity. This discontinuity, though it indeed thus fits the metal for one special stage in its mechanical preparation, yet injures it for its true use, and this is no doubt true even if the sulphur present should fail to cause redshortness. Thus

Brinell found that even with 0.56 per cent. of sulphur and 0.46 per cent. of carbon, steel was not redshort if it contained an abundance of manganese, 1.06 per cent.[1] This corresponds to about 1.52 per cent. of sulphide.

As an excuse for using such material for bolts it may be urged that, in rolling it into the long thin bolt-rods, the slag is drawn out into threads so thin that the small cross-section which they offer to the stress along the length of the bolt has only a slight weakening effect compared with the saving in the cost of machining effected by these same threads, which because they are long and thin are very effective in embrittling the chip turned off by the cutting tools. Further that it is more economical to use a slightly larger bolt, the manufacture of which is thus cheapened, than to use a smaller bolt of more expensive steel free from sulphide. What is true of bolts is probably even more true of screws, the holding power of which depends on the strength of the wood in which they are embedded, and is far less than that of the screw itself, even if made with a great excess of sulphide.

One is reluctant to believe the reports that this material has been used for the steering posts of motor cars. Such practice should be a felony.

834. To Sum Up.—As between different wrought irons, because all have enough slag to form important longitudinal weak surfaces, the tendency to break with a fibrous fracture is rather to be referred to trans-crystalline strength and ductility, and hence is to be interpreted as a merit.

Though as between different steels fiber is strictly speaking a defect, because it is an evidence of heterogeneousness, yet when we remember that it is typical of some of the very best steels, the nickel and chrome-nickel ones, and that in spite of it these steels excel most others, we see that we must be cautious in objecting to it. Here it is indeed the seamy side of what is otherwise admirable. We may almost say that such steel is likely to have the defects of its qualities, or rather that it may well have the qualities of its defects.

As between different states of the same steel or wrought iron due to heat treatment, the tendency to break with a fibrous fracture is to be regarded as a merit, because it is referable to a trans-crystalline strengthening of the metal itself, relatively to the slag and metal contacts, the strength of which may be regarded as constant.

It is probable that slag increases the flexibility, and in this sense decreases the tendency of rupture to strike across the metal rather than to shear along it. Flexibility and ductility are so interrelated that slag may possibly favor ductility, at least of certain forms and under certain conditions.

[1] "Brinell's Method of Determining Hardness," Wahlberg, *Journ. Iron and Steel Inst.*, 1901, No. II, vol. 40, pp. 237 and 254.

APPENDIX I

CERTAIN CRYSTALLINE INTERSECTIONS

835. Introduction.—Because one chief reason for examining these intersections is the wish to determine, by means of them, the crystallographic directions of the Neumann lamellæ and slipping planes, and because the data in Figs. 38 to 40, p. 270, concerning those planes give the directions of their traces on the cubic face and on the truncations of a cube by the octahedron and by the rhombic-dodecahedron, it is on this face and on these truncations that we will here seek these intersections. By the slipping planes I mean those along which the slip which is supposed to be the cause of the slip bands occurs.

836. Intersections of Cubic and Octahedral Planes with a Cubic Face and with a Cubic Truncation of an Octahedron.—If *abcd* in Fig. 43 is supposed to be the face *abcd* of Fig. 42 somewhat enlarged, it is part of a cubic plane, because it is parallel to the front and rear faces of an imaginary cube related to this octahedron as every cube must be related to every octahedron. It is not, strictly speaking, a cube face, because the sides of such a face should be parallel to the sides of the squares which make up whole cubes, such as the front and rear faces of the cube of Fig. 42. The lines *wtt'w'*, Fig. 43, form such a cube face.

abcd is the truncation of an octahedron by a cube, because it is the part of such an octahedron left bare by the slicing off of its front apex by the front face of the cube of Fig. 42.

On *abcd* of Fig. 43 the horizontal and vertical lines are the intersections of cubic planes on this cubic truncation of an octahedron, because, as we see on referring back to Fig. 42, they are such intersections as planes parallel to the top and bottom and right- and left-hand sides of this cube would make on *abcd*.

ab, *bc*, *cd*, and *da* of Fig. 43 are intersections of this cubic truncation of an octahedron by the four front planes of this octahedron. Hence the inclined dotted lines are the intersections of octahedral planes with this truncation, because they are the intersections which planes parallel to these front planes of this octahedron would here make. Thus they are the intersections of this truncation by the surfaces of a nest of parallel concentric octahedra smaller than the one here shown.

In short, while on the truncation of an octahedron by a cube the cubic intersections are at right angles to each other and at 45° with the sides of the truncation but parallel to the sides of the cube face, the octahedral intersections, though also normal to each other, are parallel to the sides of the trunca-

tion but make an angle of 45° with the cubic intersections and with the sides of a cube face.

837. Intersections of Cubic and Octahedral Planes with an Octahedral Face and with the Octahedral Truncation of a Cube.—If the six broad lines of Fig. 46, *cd*, *dh*, *hg*, *gj*, *je*, and *ec*, represent the boundary of the face *cdhgjec* of Fig. 42, they surround an octahedral plane. If they are produced till they meet, they form two equilateral triangles. *gh* is the intersection of the bottom of the cube with this plane. Hence the broken lines parallel to *gh* are intersections of like cubic planes with this plane. By like reasoning *cd* and *ej* are the intersections of the front and right-hand face of this cube with this same octahedral plane, and hence the inclined broken lines in Fig. 46 as a whole are the intersections of this octahedral plane by a nest of concentric parallel cubes.

Again referring back to Fig. 42, the upper right-hand octahedral face *cbnn'fe* intersects this lower right-hand front octahedral plane along *ce*. Hence in Fig. 46 the least inclined of the dotted lines, because they are parallel to *ce*, are the intersections of this same octahedral face by the right-hand upper faces of a concentric nest of smaller octahedra. By like reasoning the lines *dh* and *gj* in Fig. 46 are the intersections of this octahedral plane by the lower left-hand front plane and by the lower right-hand rear plane of this same octahedron. Hence the two inclined sets of dotted lines in Fig. 46 are the intersections of this octahedral plane by the front lower left-hand and rear lower right-hand faces of a nest of concentric parallel octahedra. In short, the three sets of dotted lines are the octahedral intersections of this octahedral plane.

Each of these octahedral intersections is parallel to one of the three sets of cubic intersections, so that the directions of the cubic and octahedral intersections of an octahedral plane coincide.

The octahedron of Fig. 42 truncates the eight corners of the cube. A somewhat smaller cube, such as that shown in Fig. 45, is more completely truncated by this same octahedron, so that the sides which were *cd*, *hg*, and *je* in Fig. 42 here are lengthened out to qq'', $q''q'$, and $q'q$, and the sides which were *dh*, *gj*, and *ec* are shortened down to nothing, as is seen by comparing $qq'q''$ in Figs. 45 and 46. Thus $qq'q''$ is the truncation of the cube by the octahedron. If we suppose that it is developed by swinging it about qq' into the plane of the paper, it becomes face *B* of Figs. 38 to 40. In this octahedral truncation, quite as in the octahedral plane *cdhgje*, both the cubic and octahedral intersections are parallel to the sides of the triangle.

838. Relations of the Cube and the {211} Trapezohedron.—In Fig. 119 *vlt* represents a 211 plane, cutting the top of the cubic diagonally at *lv* and the vertical axis at *n*, and therefore by definition at unity as regards this specific cube. This plane cuts the horizontal axes at m' and m'', that is, at a distance of 2 from the center, *s*, and hence complies with its index, 211, the reciprocal of the ratio between these three intercepts, 1:2:2. The intersections of these three cube faces by any other 211 plane parallel to this one

will be parallel to *lv*, *vt*, and *tl*, that is, parallel either to a diagonal or to a line drawn from the corner of a face to the middle of an opposite side.

Moreover, the intersections of planes parallel to any other planes of the {211} trapezohedron will on one face of the cube be parallel to a diagonal and on two others will be parallel to a line drawn from a corner to the middle of an opposite edge. Hence on any cubic face or plane there are six directions for these intersections, two running diagonally and four[1] running at 26° 34′ and 63° 26′, as shown in Fig. 44.

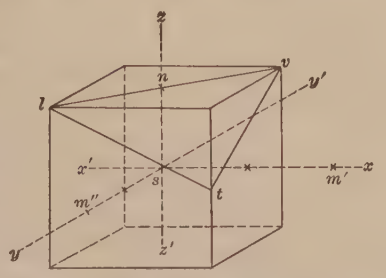

FIG. 119.—Intersections of a cube by a 211-trapezohedral plane, *lnvt*.

839. Relations of the Octahedron and the {211} Trapezohedron (Fig. 120).—We have seen that the intercepts of the three crystalline axes by any octahedral plane are equal, and that those made by a 211 trapezohedral plane are in the ratio 1 : 2 : 2. Hence planes *BDC*, drawn so that *DO* = *BO* = *CO* = 1, and *EFG* drawn so that *EO* = *FO* = *GO* are both octahedral, and *ABC*, drawn so that *BO* = *CO* = 1 and *AO* = ½ is a 211 plane. Clearly the intersection of *ABC* and *DBC* at *BC* is parallel to an edge of the octahedral face, because it is such an edge itself. Equally clearly the intersection *HI* of *ABC* with any smaller octahedral plane, such as *EFG*, also is parallel to *BC* and hence is parallel to an edge of an octahedral face.

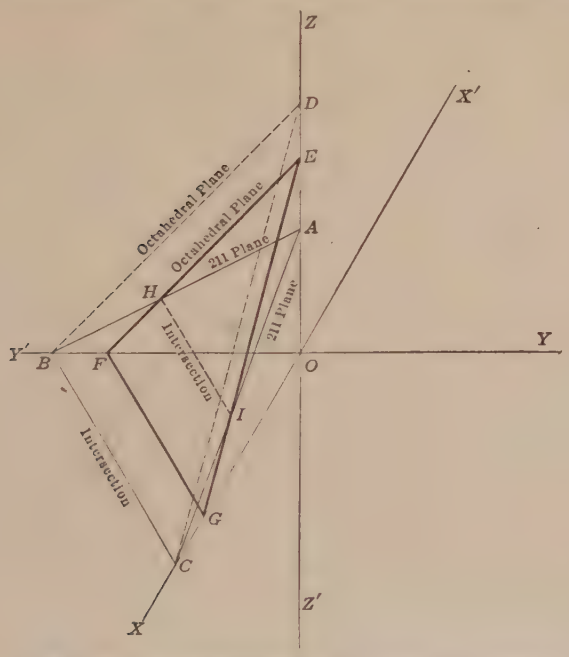

FIG. 120.—An intersection, *HI*, of a 211 trapezohedral plane by an octahedral plane is parallel to one of the sides, *FG*, of the corresponding octahedral face.

This conclusion aids our interpretation of the slip planes (see §447).

Like reasoning shows that the intersection of a 211 plane which cuts axes *x* and *z* at unity and *y* at ½, with this same octahedral plane *EFG*, is parallel to *EG*; and that the intersection of still a third 211 plane is parallel to *EF*; so that the 211 planes collectively can cause intersections parallel to the three sides of an octahedral face.

NOTE TO FIG. 120.—The intercepts of *ABC* are ½, 1, and 1; hence it is a 211 plane. The intercepts of *DBC* are 1, 1, and 1; hence it is an octahedral plane. The intercepts of *EFG* are 0.78, 0.78, and 0.78; hence it is an octahedral plane.

[1] Four, not eight. The 26° 34′–63° 26′ lines starting from any one corner are parallel to those starting from the corner diagonally opposite.

840. Intersections of Cubic Planes with a Rhombo-dodecahedral Truncation of a Cube, Fig. 48 and C of Figs. 38 to 40.—Such a plane as *LMNP*, Fig. 121, parallel to plane *ACFH* running diagonally across the cube, is clearly a rhombo-dodecahedral plane, because, while parallel to the vertical axis *zz'*, it cuts the horizontal axes *xx'* and *yy'* at one and the same distance, zero in the case of *ACFH*, and hence it has the ratio $\infty : 1 : 1$ between its intercepts, and hence it has the index 011 of the rhombo-dodecahedral planes. This figure *LMNP* is clearly a truncation of this cube by a rhombo-dodecahedral plane. It is a rectangle which becomes a square when $DM = DC \div \sqrt{2}$. Its sides are the intersections of the top, bottom, front, and left-hand faces of the cube on this truncation. Moreover, the rear and right-hand faces of this cube if produced would trace on this same plane lines parallel to *LP* and *MN*. In short the intersections of this rhombo-dodecahedral truncation of a cube by planes parallel to the six faces of a cube, that is by cubic faces, follow in all only two directions, one parallel to *LM* and the other to *MN*. Moreover, the intersection of this plane with the top of the cube is parallel to the diagonal of that top.

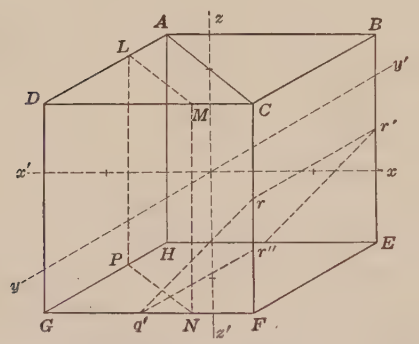

FIG. 121.—Intersections of a cube by rhombo-dodecahedral planes, *LMNP* and *q'rr'r''*.

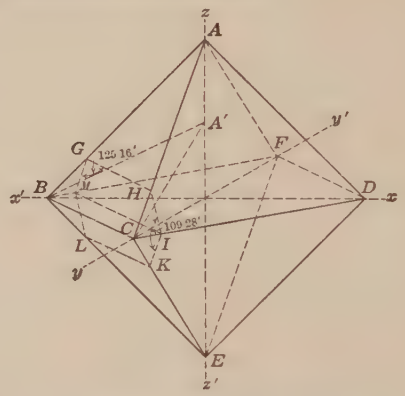

FIG. 122.—Intersection of an octahedron by a rhombo-dodecahedral plane, *GHIKLM*.

An altogether like rhombo-dodecahedral truncation would be made by the plane *C* or *q'rr'r''* of Fig. 121. If we imagine that this truncation is swung about its intersection *q'r* into the plane of the paper, its relation to the face *DCFG* and to the cube proper would be exactly those which the correspondingly lettered rectangle *q'rr'r''* of Figs. 38 to 40 bears to the cube face *A* and to the cube to which that face belongs. The directions of the cubic intersections would be *uu'* and *vv'* of face *C* of Fig. 38, of which *vv'* is parallel to the cubic intersection *q'r* of this face, and the other is normal to this first. This same face *C* is reproduced in Fig. 48 which therefore gives the directions, 1 and 2, of the cubic intersections of a rhombo-dodecahedral truncation of a cube. These directions, then, *vv'* and *uu'* in face *C* and lines 1 and 2 in Fig. 48, are all the possible intersections of these rhombo-dodecahedral truncations of the cube by the cubic planes.

841. Intersections of Octahedral Planes with a Rhombo-dodecahedral Truncation Plane of a Cube, C of Figs. 38 to 40, and Fig. 49.—The plane

GHIKLM of Fig. 122, parallel to its vertical axis *zz'* and also parallel to the edges *BC* and *DF* of the octahedron, is clearly a rhombo-dodecahedral plane, because, while its intercept on the vertical axis is ∞, those on the horizontal axes are equal *inter se*, for this plane is parallel to *BC*, the intercepts of which on those axes are by definition unity and hence equal. If, now, we regard those equal intercepts of *GHIKLM* on those axes as unity, the three intercepts of that plane are ∞, 1, and 1, so that its index is 011, which is the rhombo-dodecahedral index.

It can be shown that the internal angle *HIK* = *GML*, or in general the angle between the sides of the octahedron on any such rhombo-dodecahedral plane, is 109° 28′, and that the angle *GHI* = *HGM* is 125° 16′.

GH, *HI*, and *IK* are the intersections of three faces of the octahedron on the rhombo-dodecahedral plane *GHIKLM*, and hence these angles 109° 28′ and 125° 16′ are the ones between the octahedral intersections of any such plane, such as the rhombo-dodecahedral truncation of a cube, C of Figs. 38 to 40.

If we now conceive that this rhombo-dodecahedral plane, after being intersected at *GHIKLM* by the octahedron, is moved parallel to itself so that it coincides with plane *LMNP* of Fig. 121, *GH* of Fig. 122 must be parallel to *LM* of Fig. 121, which is a cubic intersection of such a truncation, first, because both are in the same vertical plane, second, because each is horizontal, being parallel to the horizontal plane passing through the axes *xx'* and *yy'*, and third, because any lines in one plane and parallel to a plane not parallel to theirs must be parallel to each other. Hence on the rhombo-dodecahedral truncation of a cube, such as *LMNP* of Fig. 121, or as face *C* of Figs. 38 to 40, or as Fig. 49 which we draw as the equivalent of these *C* faces, one of the octahedral intersections will be parallel to a cubic intersection *LM* of that truncation, such as *q'r* is in each of these figures, and the two other octahedral intersections must make with this first this angle 125° 16′, and with each other the angle 109° 28′, as lines 1, 2, and 3 do in Fig. 49.

Returning now to Fig. 122, the three directions *GH*, *HI*, and *GM* are the only possible directions which the intersections of any octahedral planes can make on a rhombo-dodecahedral plane, as is seen on inspection. Thus the planes *ABF* and *BFE* are parallel to *CDE* and *CAD* respectively, and hence their intersections *GM* and *ML* with the 011 plane are parallel to *IK* and *HI* respectively. The reason for which *GH* is parallel to *BC* and to a cubic intersection of the 011 plane applies also to *LK* the intersection of *BCE*; and there remain only *ADF* and *FDE*, which, because they are parallel to *BCE* and *ABC*, must intersect the 011 plane parallel to *GH* and *KL*.

In short, the directions 1, 2, and 3 of Fig. 49 are the only possible directions of the intersections which octahedral planes make on a rhombo-dodecahedral truncation of a cube.

By like reasoning the numbered lines in Figs. 47 and 50 can be shown to be the intersections of the planes of a {211} trapezohedron on an octahedral and on a rhombo-dodecahedral truncation of a cube, such as faces *B* and *C* of Figs. 38 to 40 are.

APPENDIX II

842. Miscellaneous Tables.

TABLE 34.—INCREASE OF THE BRINELL AND SHORE HARDNESS WITH THE CARBON CONTENT
Based chiefly on the data of Brinell and of Shore

No.	Carbon content, per cent.	Shore scleroscope hardness				Brinell or ball hardness			Ratio: Brinell ÷ Shore hardness	
		Condition of the metal				Condition of the metal				
		Annealed A	Normal merchantable B	Hardened C	Ratio $\frac{C}{A}$	Annealed D	Hardened E	Ratio $\frac{E}{D}$	Annealed	Hardened
1	2.04[c]	.....	.....	.....	.....	286[c]	Cracked	.....	.....	.....
2	1.75	40–45	50–55	110	2.44–2.75	.....	.....	.....	.....	.....
3	1.72[d]	.....	.....	.....	.....	286[d]	477[d]	1.67	.....	.....
4	1.50	37–40	47–50	110	2.75–2.97	.....	.....	.....	.....	.....
5	1.25	35	44	110	2.78	262[f]	627[f]	2.39	7.5	5.7
6	1.00	30–35	40	107	3.06–3.57	259[a]	627[a]	2.42	7.4–8.6	5.9
7	0.80	28–30	38	105	3.50–3.75	235[a]	648[a]	2.76	6.7–7.8	6.2
8	0.66	.....	.....	.....	.....	202[b]	578[b]	2.86	.....	.....
9	0.65	.....	.....	.....	.....	235[f]	652[f]	2.77	.....	.....
10	0.60	27	36	102	3.78	225[a]	628[a]	2.79	8.3	6.2
11	0.50	26	34	96	3.69	204[a]	579[a]	2.84	7.8	6.0
12	0.45	.....	.....	.....	.....	194[f]	555[f]	2.86	.....	.....
13	0.40	25	32	90	3.60	175[a]	478[a]	2.73	7.0	5.3
14	0.35	24	30	82	3.42	156[f]	402[f]	2.58	6.5	4.9
15	0.30	23	28	70	3.05	.....	.....	.....	.....	.....
16	0.25	22	27	58	2.64	143[f]	311[f]	2.11	6.5	5.4
17	0.20	21	26	50	2.38	115[f]	196[f]	1.70	5.5	3.4
18	0.15	20	25	43	2.15	.....	.....	.....	.....	.....
19	0.10	19	24	35	1.84	97[f]	149[f]	1.54	5.1	4.4
20	0.05	18	23	25	1.39	.....	.....	.....	.....	.....
21	0.01[e]	.....	.....	.....	.....	75[e]	107–112	1.46	.....	.....

Note to Table 34.—The Shore data have been kindly given by Mr. Shore as representing his experience.

The Brinell hardness data comprise: first, his own data, marked [f] and [b] (*Journ. Iron and Steel Inst.*, 1901, No. I, vol. 59, p. 267, Table IX); second, data derived by interpolation from his, marked [a]; and third, direct tests on certain of my steels, marked [c], [d], and [e], and forming Nos. 1, 3, and 21 of this table. In deriving the interpolated data the steel of 0.66 per cent. of carbon was neglected, because of its large manganese content. The composition of these steels is given in part in Table 35 and in part in Table 39.

[c] Steel No. 25.1 of Table 39.

[d] Steel No. 25 of Table 39.

[e] Steel No. 0 of Table 39.

TABLE 35.—COMPOSITION OF THE STEELS USED BY BRINELL AND SHORE IN THE HARDNESS EXPERIMENTS FROM WHICH THEIR DATA IN TABLE 34 ARE DERIVED[a]

For Shore hardness					For Brinell hardness				
C	Si	Mn	P	S	C	Si	Mn	P	S
1.34	0.216	0.167	Low	Low	1.25	0.60	0.20	0.027	0.010
1.28	0.261	0.218	Low	Low	0.92	0.28	0.25	0.026	0.012
1.24	0.141	0.214	0.015	0.023	0.78	0.37	0.20	0.028	0.011
1.10	0.258	0.214	Low	Low	0.66	0.33	0.18	0.028	0.010
0.95	0.141	0.354	0.018	0.020	0.65	0.27	0.49	0.028	0.011
0.95	0.251	0.228	Low	Low	0.45	0.27	0.45	0.028	0.018
0.88	0.136	0.252	0.029	0.031	0.35	0.26	0.49	0.027	0.015
0.75	0.156	0.241	0.024	0.022	0.25	0.30	0.41	0.028	0.012
0.62	0.198	0.207	0.019	0.023	0.20	0.018	0.41	0.027	0.015
.....	.....	.....	.....	.....	0.10	0.007	0.10	0.026	0.020

[a] The composition of my steels, Nos. 1, 3, and 21 of Table 34, is given in Table 39.

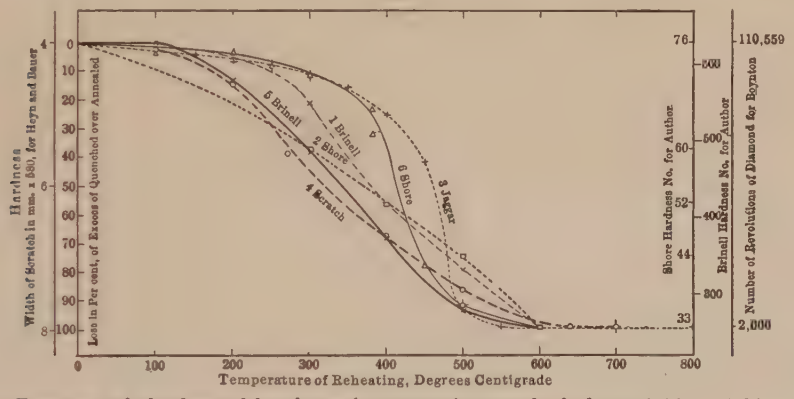
Hardness
Width of Scratch in mm. x 880, for Heyn and Bauer
Loss in Per cent. of Excess of Quenched over Annealed
1 Brinell
2 Shore
4 Scratch
5 Brinell
6 Shore
8 Jaggar
Shore Hardness No. for Author
Brinell Hardness No. for Author
Number of Revolutions of Diamond for Boynton
Temperature of Reheating, Degrees Centigrade

FIG. 123.—Progress of the loss of hardness in tempering steel of about 0.80 to 1.20 per cent. of carbon.

NOTES TO FIG. 123.—1. *Brinell*, 2. *Shore:* THE AUTHOR. Short flat bars, ½ in. × ⅜ in., of 1 per cent. carbon steel were quenched from about 1,000°, and then set in the cold furnace. They were then raised severally along with the furnace to the temperatures indicated, at which they were held for 10 minutes.

3. *Jaggar:* BOYNTON (*Journ. Iron and Steel Inst.*, 1906, No. II, vol. 70, p. 316). Pieces of a 0.86 carbon steel, ½ in. long and ½ in. in diameter, were quenched from 850°. A single specimen was heated successively to the several temperatures indicated, kept ½ hour at each, and re-ground after each determination. The 500° determination was checked by heating another piece to this temperature, with concordant results. The Jaggar scleroscope measures the hardness by the number of revolutions of a diamond point under constant load and speed, needed to drill to a depth of 0.01 mm.

40. *Scratch:* HEYN AND BAUER (*Stahl u. Eisen*, 1906, vol. 26, p. 780). Pieces 1 in. × 1 in. × ¼ in. of 0.95 C steel were quenched from 900°. They were reheated and held 3 hours at the several temperatures. The hardness was determined by the width of a scratch made by a diamond under constant load.

5. *Brinell*, 6. *Shore:* MAURER (*Rev. de Metallurgie, Mem.*, 1908, vol. 5, p. 711, and *Metallurgie*, 1909, vol. 6, p. 39). Steels of 0.83 and 1.20 carbon were quenched from 800°.

In these notes the italicized names following the designation number (*Brinell*, *Shore*) refer to the method of determining hardness, the names in small capitals (THE AUTHOR, BOYNTON) to the observer.

TABLE 36.—BRINELL AND SHORE HARDNESS OF SOME SPECIAL STEELS. (Cf. pp. 465–467)

No.	Authority	Chemical composition, per cent.								State	Hardness
		C	Al	Cr	Mn	Ni	Si	V	W		
						Brinell hardness					
1	Guillet	0.15		0.06		2.13				Annealed	131
2	"	0.34		10.25		1.9				"	579
3	"	0.07		0		3.89				"	118
4	"	0.27		8.26		4.01				"	500
5	"	0.80				5.0					250
6	"	0.80				7.0					525
7	"	0.80				15.0					150
8	"	0.80				30.0					175
9	"	0.80			3.08						110
10	"	0.80			7.20					As rolled or forged	445
11	"	0.80			14.4					(bruts de forge)	210
12	"	0.80			33.4						140
13	"	0.80		0.5–11.5							300–390
14	"	0.80		14.5							470
15	"	0.80		26.5							170
16	"	0.80							2.7–10	Annealed	300±
17	"	0.80							14.7	"	350
18	"	0.80							19.2	"	280
19	"	0.80							26.2	"	352
20	"	0.80						0.8		"	330
21	"	0.80						10.25		"	180
22	"	0.80					2.0			"	277
23	"	0.20					5.1			"	248
24	"	0.80	0.5–15.0							"	210–277
25	Sauveur	1.25			12.50					Austenite, quenched from 1,100°	177
26	"	1.25			12.50					Quenched from 1,100°, reheated 575°, 90 hours, f. c.	356
27	Hopkinson & Hadfield	1.21			12.36		0.22			Austenitic, quenched from yellow heat	207
28	" "	1.21			12.36		0.22			Quenched from yellow heat, reheated 550°—48 hours	495
29	Carpenter and Hadfield and Longmuir	0.47			0.95	Nil					202
30		0.40			0.82	4.25					217
31		0.42			1.03	4.95					321
32		0.52			0.92	6.42					532
33		0.43			0.79	7.95					578
34		0.41			0.85	12.22					555
35		0.45			0.83	15.98					293
36		0.41			0.96	19.91					131
						Shore hardness					
37	Maurer									Austenite quenched from 1,050°	17
38	"	1.94			2.2		0.98			Quenched from 1,050°, reheated to 400°	28
39	"									Quenched from 1,050°, reheated to 950°	21
40	Roush	Less than 0.10	Remelted electrolytic iron-nickel alloys		0.25–4.0					Annealed at 900°	14–14½
41	"				13.0–22.0					" " "	31–34½
42	"				35.0–45.0					" " "	15–17

Notes to Table 36.—1 to 4. Guillet, *Comptes rendus*, 1914, vol. 158, p. 413. He gives also the hardness for intermediate chromium contents.

5 to 8. Guillet, *Alliages métalliques*, Dunot and Pinat, Paris, 1906, p. 289.

9 to 12. *Ibid.*, p. 320.

13 to 15. *Ibid.*, p. 329.

16 to 19. *Ibid.*, p. 337. He gives also results of Böhler, which show hardness of 0.85 to 1.44 for unquenched and of 2.10 to 3.39 (?) for quenched W steels with 0.66 to 1.16 C and 0.46 to 7.78 W.

20 to 21. *Ibid.*, p. 351.

22 to 23. *Ibid.*, p. 359.

24. *Ibid.*, p. 368. Guillet gives many additional data. For this series with 0.80 per cent. of carbon he gives the hardness for contents of most of the alloying metals intermediate between those here given. He gives also an additional series of like alloy steels, with either 0.15 or 0.20 per cent. of carbon.

25 to 26. Sauveur, *Trans. Amer. Inst. Min. Eng.*, 1914, vol. 50, p. 510.

27 to 28. Hopkinson and Hadfield, *idem*, p. 482. The steel contained 0.080 per cent. P and 0.030 per cent. S.

29 to 36. Carpenter, Hadfield and Longmuir, *Seventh Report Alloys Research Comm.*, excerpt *Proc. Inst. Mech. Eng.*, Nov., 1905, p. 880.

37 to 39. Maurer, *Rev. Metallurgie, Memoires*, 1908, vol. 5, p. 726; *Metallurgie*, 1909, vol. 6, p. 41. A home-made scleroscope was used.

40 to 42. Roush, "Hardness Tests of Alloys of Nickel and Copper with Electrolytic Iron," *Bull. Univ. Wisconsin; cf. Metall. and Chem. Engin.*, 1910, vol. 8, p. 469.

Table 37.—Reagents for Detecting Phosphorus Banding

	1 Heyn	2 Rosenhain and Haughton	3 Stead	4 Le Chatelier and Lemoine	5 Brearley	6 Whiteley
Ammonio-cupric chloride, grams	10	.....	.....	.....	.....	.....
Cupric chloride, "	.....	10	10	10	.....	.....
Cupric oxide, "	.....	.....	.....	.....	.....	0.04
Ferric chloride, "	.....	30	.....	.....	2.5	.....
Stannous chloride, "	.....	0.5	.....	.....	.....	.....
Magnesium chloride, "	.....	.....	40	40	.....	.....
Hydrochloric acid, conc., c.c.	.....	100	20	20	2.5	.....
Water, "	120	1,000	100±	180	1,000	.....
Nitric acid, "	.....	.....	.....	.....	.....	6
Alcohol, "	.....	.....	1,000	1,000	.....	194

Notes to Table 37.—1. Heyn and Bauer, *Metallographie*, Göschen, Leipsic, 1909, vol. 1, p. 21; Martens-Heyn, "Materialenkunde," Springer, Berlin, 1912, Vol. II, A, p. 178. The specimen is immersed in the reagent for 1 minute; the deposited copper is then removed by gentle dabbing, not rubbing, under water, and the specimen is dried as usual.

2. *Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, p. 515. The polishing should be finished on a very wet block, and the specimen washed with absolute alcohol and immersed in the reagent for 10 seconds to 2 minutes. The copper adheres firmly and must not be removed. Pearlitic areas remain bright; the ferrite is darkened by the deposited copper, the purest parts darkening the fastest. Quenched steels give the usual martensitic markings. Pitting can be avoided by diluting the reagent.

3. *Proc. Cleveland Inst. Eng.*, 1914–1915, p. 78; *Journ. Iron and Steel Inst.*, 1915, No. I, vol. 91, p. 173. The reagent is applied drop by drop till the structure is developed. The specimen is then washed with a jet of boiling water, and then with alcohol, which evaporates soon. The indications are similar to those given by Rosenhain and Haughton's reagent. The quantity of water to be used is the least possible.

4. *Comptes rendus*, 1915, vol. 161, p. 373 (*Journ. Iron and Steel Inst.*, 1915, No. I, vol. 91, p. 196). The reagent may be applied like Stead's, No. 3. The contrast can be increased by electrolysis, the specimen being connected with the positive pole of a single storage battery cell and immersed in the liquid. By keeping the other electrode 2 cm. away, the current is kept below 50 milliamperes. See also *Rev. Metallurgie, Mem.*, 1915, vol. 12, p. 649.

5. Discussion of Rosenhain and Haughton's paper, *Journ. Iron and Steel Inst.*, 1914, No. I, vol. 89, p. 525. Also cited by Stead. The hydrochloric acid may be omitted. With ordinary mild steels the neutral solution colors the ferrite only at first, but on etching for a longer time the pearlite becomes dark and the pattern is reversed.

6. Cited by Stead (*op. cit.*). The reagent is applied for 1 minute until a brown or purple film forms. After washing, this film is easily removed with the finger. On lightly repolishing on cloth, the phosphoric areas appear in relief when examined with a hand lens. The etching may be repeated.

Of these reagents, that of Stead, and more especially that of Le Chatelier, give very striking contrasts.

APPENDIX II (*Continued*)

843. Description of the Materials Experimented with for this Work, Shown in the Micrographs of Plates 1 to 45.

The chemical composition, tensile properties, and origin of the materials experimented with are given in Table 39.

Note 1. *Frontispiece.*—The molten metal probably did not contain more than 0.06 of either phosphorus or sulphur, so that there has been considerable enrichment in both elements

	C	Si	Mn	P	S
Composition of a crystal	0.43	0.191	1.05	0.098	0.101
Metal near crystal, about 1 in. from the cavity	0.345	0.16	1.02	0.076	0.080

Note 2. *Fig. E of Plate* 2.—The crystal is 39 cm. (15.3 in.) long. It was found projecting downward into the cavity in the sinking head of a cylindrical 100-ton ingot of soft open-hearth steel. The sinking head itself was 1 meter (39.3 in.) in diameter and 1½ meters (59 in.) high.

Note 3. *Fig. O of Plate* 3.—The cast iron contained

Comb. C	Graph.	Total C	Si	Mn	P	S
4.271	0.255	4.526	0	0.04	0.031	0.002

The content of phosphorus, sulphur, and manganese is that of the pure Swedish wrought iron from which this cast iron was made by melting it in a crucible brasqued with sugar charcoal, and hence is approximately that of the cast iron itself.

Note 4. *Fig. D, Plate* 4.—This extremely pure cast iron was made by melting Burgess' electrolytic iron in graphite crucibles brasqued with a mixture of sugar carbon and vaseline, together with a very small quantity of silicon introduced intentionally. It was re-melted at about 1,350° in an Acheson graphite crucible covered with Acheson graphite, almost free from ash, and allowed to solidify very slowly in the furnace. About 4 hours was occupied in cooling to 700°.

Note 5. *Fig. A of Plate* 6. The cast iron contained

Comb. C	Graph.	Total C	Si	Mn	P	S
1.598	1.356	2.954	0.040	nil	0.050	0.035

Note 6. *Figs. A to E, Plate* 8.—The five Figs. A to E represent five wads of steel (0.40 per cent. carbon) ⅝ in. in diameter and about ½ in. long, which were heated in an atmosphere rich in carbonic oxide, in the closed porcelain tube of a Heraeus furnace, to 1,000° for 10 minutes, cooled slowly in the furnace to the temperatures indicated, and then quenched in cold water, except E which was cooled slowly to room temperature. A, B, and C were treated in one operation, and D and E in a second. In the interval between quenching any given specimen and quenching the preceding, the thermo couple lay exactly above that specimen. (Author, V, p. 308.)

Note 7. *Figs. F to H of Plate* 8.—A bar of this Steel 14 of 0.92 carbon, ¼ in. square and 2 in. long, was coarsened by heating to 1,200°±, and furnace cooling slowly. Then one end was heated well above the transformation range by immersing it in a bath of molten potassium sulphate, while the other end projected up into the air, thus establishing a rather flat thermal gradient. While in this condition the bar was drawn and quenched in water. Thus Figs. F, G, and H show the structure at three points, one below and two above the transformation range at the moment of quenching, as modified by so much of the transformation as occurred during the quenching itself.

Note 8.—*Figs. I to N of Plate* 8 show the structure at various points in a single bar of Steel I (carbon 0.40 per cent.), at a series of points successively lower in the transformation range. The bar was first heated nearly evenly within an iron tube in a gas forge to about 1,325°. The heating up occupied about 20 minutes. It was then held at this high temperature for about 35 minutes, so as to coarsen the austenite structure. The tube and bar were then drawn forward slowly, through a small orifice in the forge door, till about half their length protruded into the air, the temperature within the forge falling slowly the while. This slow forth-drawing, which occupied about 15 minutes, caused the cooling of the outer end to outrun that of the middle and rear of the bar, and so established a flat thermal gradient along the length of the bar, passing through the transformation range and part of Regions 4 and 8A. Before quenching, the bar was left in this position for about 15 minutes, in order that the several points in it, which at the moment of quenching were in the transformation range, should have reached their several positions in that range by means of a cooling slow enough to build up the cellular structure, yet not so slow as to lead to its degeneration and break-up. The bar was then drawn from the tube and quenched in water, in order to fix approximately the structure thus developed.

Thus the two series of specimens, Figs. A to E and I to N, are of the same steel after treatment which was in general alike, save that the specimens shown in Figs. I to N underwent a retarded air-cooling, which was very much less slow than the furnace cooling of the specimens shown in Figs. A to E, the faster cooling preventing and the slower cooling allowing the break-up of the network. Beyond this, each of the latter specimens was a separate piece, drawn and quenched separately when it had cooled as far as its individual quenching temperature. (Author, V, pp. 295 and 308.)

The slower cooling in Figs. A to E has given such an opportunity for the coalescence of the precipitated ferrite that this in Figs. C to E has effectively masked the network structure. The more rapid cooling in Figs. I to N has restrained the coalescence of the ferrite, and thus prevented that coalescence from masking the network structure. The white rims around the ferrite network of Figs. M and N suggest undercooling past the eutectoid composition and the precipitation of cementite on the breaking up of the undercooling, thus causing these white borders. But closer examination does not confirm this interpretation. In particular these light rims do not darken on etching with sodium picrate.

Note 9. *Figs. O to Q of Plate* 8.—Wads of this steel about $\frac{5}{8}$ in. in diameter and 0.5 in. long were heated to 1,000°, held 10 minutes, cooled within the furnace to the quenching temperature given below the several figures, held there for 30 minutes, drawn, and quenched in cold water. The progressive increase in the quantity of white areas from Fig. O to Q represents the increase of the quantity of ferrite precipitated in cooling from 800°, Fig. O, to 780°, Fig. Q. All of the white matter shown in each of these figures must have been present at the moment of quenching, because any additional quantity of ferrite which precipitated during the quenching would have no time to coalesce into masses visible with this magnification.

Note 10.—*Figs. R to Y of Plate* 8, *the course of re-absorption with steadily rising temperature,* represent a series of points in the length of one and the same piece of steel, No. 9, of 0.40 per cent. of carbon, after a differential treatment which completed the re-absorption of pro-eutectoid ferrite in one end of the piece without allowing it to begin in the other end. In order to bring this about the piece was overheated to 1,350° and cooled slowly so as to assemble the ferrite together into coarse compact masses easy to recognize. Fig. R shows the structure of the coarsened bar at this stage. Successive stages in the re-absorption of the ferrite thus massed were reached by heating the bar differentially so as to create a flat thermal gradient along its length, running through the transformation range. This was done by immersing one end in a bath of molten potassium sulphate and allowing the other to project into the air, under a suitable cover. These stages were then fixed by quenching the piece in cold water, and they were then revealed by etching the bar and photographing a series of fields along its length, with the results shown.

Fig. S shows, in its left-hand part, a polygonal grain which had not passed above $Ac1$, and

therefore is in the condition of ferrite plus pearlite, of which the former appears as the light network and spines, the latter as the dark ground mass. The light-colored polygons at the right of this dark one had passed above *Ac*1, and therefore at the moment of quenching consisted of the unreabsorbed ferrite, shown still in white spines and network, and the austenite into which the pearlite had changed, here appearing as a martensite ground mass which, though slightly less light than the ferrite, is yet far lighter than the pearlite in the left-hand grain.

The successive figures T to V show stages in the further re-absorption of the ferrite. First the interior or cleavage ferrite disappears, then the network becomes discontinuous, and in Fig. V disappears. In some parts the network which the ferrite had formed can be traced beyond where the ferrite itself vanishes, but this darkening seems to be only a shadow effect, because under proper illumination the bottom of the grooves which form the grain boundaries appears of the same pearly white as the interior of the grains.

The progressive increase from Fig. W to Fig. Y of the light martensite and decrease of the dark troostite represent the increasing lag of the transformation from the stage of martensite to that of troostite, an increase which the rise of the quenching temperature naturally induces. (See the Author, "Why does Lag Increase with the Temperature from which Quenching Starts?" *Trans. Amer. Inst. Mining Eng.*, 1913, vol. 45, p. 516.) The white spottings at the right-hand part of Fig. V indicate a possible cause of error. These occur above where the absorption of the ferrite network has become complete, and hence presumably well above *Ae*3. If such spots should be noticed in cooling down, they might well be taken for an early precipitation of ferrite, and thus the line *A*3 might be set not only too high but perhaps very much too high.

A careful study of the white constituent which forms the ground mass of the polygons in Figs. T, U, and V leaves no doubt that it is martensite.

One asks why this cooler and hence less lagging part of the specimen should be caught in the stage of martensite though the hotter part represented by Fig. W has transformed beyond into that of troostite, common report being that of the two it is quenching from above *A*3 that yields martensite and quenching from the lower part of the transformation range that yields troostite. Two reasons why martensite should be found in the cooler but not in the hotter parts suggest themselves: First, at the moment of immersion the austenite in the parts just above *A*1 was a higher carbon austenite than that in the hotter parts, and it is most familiar that an increase of carbon content increases lag; second, in quenching it was the hot end of the piece that entered the water first, and hence in its rapid descent through the water it was less surrounded by steam than the cooler parts above it. A repetition of this experiment gave like results.

The cleavage ferrite is often much more abundant near the boundaries of a given polygon than in the central part of it, suggesting that this cleavage ferrite has been trapped here in the course of its migration to the grain boundaries.

The shorter opportunity for balling up by surface tension explains why the ferrite network remains continuous to a further stage in its re-absorption here, in rising temperature, than when it is re-absorbed at stationary temperature. (Howe and Levy, IV, pp. 601 to 603.)

Note 11. *Figs. A to H of Plate* 9.—*Outline of procedure.* Figs. A to H show the structure of this steel after highly heating, partly cooling to a series of temperatures above, in, and below the transformation range, and then quenching. In each case the hardened steel was next "osmonditized," *i.e.*, heated to 400° so as to bring it to the osmondite stage, in order both that it might be cut readily, and that the contrast between the cementite and the ground mass might be intensified.

Preliminary or Austenitizing Heating.—The specimens were heated under borax in a porcelain crucible in an electric resistance furnace for 20 minutes to above 1,200°, with a maximum of 1,280°, or about 140° above the solidus after Gutowsky (*Metallurgie*, 1909, vol. 6, p. 739). Piece A of this row was heated separately in the same furnace to 1,200°, cooled during 6 minutes to 960° and quenched. 960° is about 65° below SE according to Gutowsky, but

only about 10° below SE after Wark (*Metallurgie*, 1911, vol. 8, p. 711), and very close to SE as determined under comparable conditions for this very steel by the Bureau of Standards, Washington (Howe III, pp. 330–332; Burgess, Crowe and Rawdon I, Bulletin, p. 605).

Cementite-generatiag Sojourn.—After having all initial cementite thus re-absorbed by the austenite in this preliminary heating, the generation of pro-eutectoid cementite was next started by transferring the specimens simultaneously into a bath of fused chlorides of potassium and sodium in a graphite crucible 4½ in. in diameter, in a vertical tubular furnace 5 in. in diameter and 22 in. long, heated externally by the resistance of a spiral of nichrome wire. The initial temperature of the salt bath was 930°. After staying in the salt for the length of time indicated below the several micrographs, at slowly falling temperature, the specimens were drawn severally and quenched in cold water. The arrangement of these two furnaces has been described in Howe and Levy I, p. 11, and Howe and Levy II, pp. 3 *et seq.*

Note 12. *Figs. I to N of Plate* 9.—The steel was coarsened by holding between 1,200° and 1,300° for about 30 minutes, and cooled slowly in the furnace. The several specimens were then heated separately for 30 minutes to the temperatures indicated below the micrographs, and quenched in water.

Note 13. *Figs. P to T of Plate* 9.—The treatment was approximately like that of specimens A to E of Plate 8, as explained in Note 6.

Note 14. *Fig. F of Plate* 12.—From the plate of a chilled cast-iron car wheel made of coke iron. Such coarse graphite flakes are not uncommon in such car wheels. The individual field here shown is one of the coarsest in this specimen. In many other fields the flakes are materially finer.

The composition of this iron is as follows:

Total C	Gr	Comb. C	Si	Mn	P	S
3.10	2.60	0.50	0.73	0.48	0.39	0.14

(Johnson [Ironkem], *Iron Age*, 1912, vol. 89, p. 1209.)

Note 15. Plate 27.—Specimen A is shown etched in Howe and Levy VI, Fig. B of Plate 3; specimens B, C, and D are shown etched in Author V, Figs. E and F, Row 6, Plate 3; specimens similar to E and F are shown etched in Howe and Levy VI, Figs. A to C, Plate 1. These prior micrographs do not show the deformation figures.

Note 16. Plate 42.—After polishing A, B, and F were notched with a fine saw or file cut. The notch was then propagated by hammering.

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y printing their source in Column 9 in capitals. Those repr

Plate	Fig.	Referred to in §	Carbon, per cent.	Material	
Col. 1	2	3	4	5	
Frontispiece		187, 343	0.043	Medium-carbon steel ingot.	As cast. S
Facing Title		195	.....	.....	As cast. . .
1	A	71, 74	0.92	Eutectoid steel.....	Heated to 9
"	B	" "	0.59	Hypo-eutectoid steel....	" " 1
"	C	" "	1.2	Hyper-eutectoid steel...	" " 1
"	D	75	0.59	Hypo-eutectoid steel....	" " 1
"	E	"	1.45	Hyper-eutectoid steel...	" " 8
"	F	81	5± est.	Spiegeleisen.....	Probably sc
"	G	81, 175	5± est.	"	"
2	A	99	1.7	Blister steel.....	Heated to 1
"	B	97	0.46	Hypo-eutectoid steel ...	" " 1
"	C	71, 74	1.14	Hyper- " " ...	" " 1
"	D	" "	1.14	Hyper- " " ...	" " 1
"	E	187	Low	Steel.....	As cast....
"	F	108, 541	0.214	Low-carbon steel.....	Heated to 1 and pulled
3	A	92, 174, 240	0.023	Electrolytic iron.....	Heated to 92
"	A2	768, 770	0.14	Low-carbon steel.....	" " 1,
"	B	92	0.235	" " "	Heated to 1,
"	C	"	0.40	Medium-carbon steel...	" "
"	D	"	0.59	" " " ...	" "
"	E	"	0.78	" " " ...	" "
"	F	"	0.92	Eutectoid steel.....	" " 1,
"	G	"	1.14	Hyper-eutectoid steel...	" " 1,
"	H	"	1.45	Very high carbon steel..	" "
"	I	92, 205	1.72	Extremely high carbon steel.	" "
"	I2	94	2.04	Eutectiferous steel.....	As rolled....
"	J	92, 205	.....	Very Hypo-eutectic white cast iron	As cast...
"	K	" "	3.055	Moderately Hypo-eutectic white cast iron	" " ...
"	L	" "	.....	Slightly Hypo-eutectic white cast iron	" " ...
"	M	92,	.....	Spiegeleisen.....	As cast.....
"	N	196,	.....	Spiegeleisen.....	" " .....
"	O	205	4.526	Hyper-eutectic cast iron.	Poured into w
4	A	118, 123	.....	Black heart malleable casting	As annealed in
"	B	" "	.....		
"	C	" "	.....		
"	D	120	.....	Gray cast iron.....	Solidified and
"	E	118	3.16	" " "	As cast, same Plate 6
"	F	124, 299	.....	Johnson's oxygenated cast iron.	As cast.....

TABLE 38.—DESCRIPTION OF THE SPECIMENS SHOWN IN THE MICROGRAPHS OF PLATES 1 TO 45.—*Continued*

Plate	Fig.	Referred to in §	Carbon per cent.	Material	Treatment	Fig.	No. in Table 39	Description and reference
Col. 1	2	3	4	5	6	7	8	9
					Plate 5. Charcoal iron, and transformation structures in cast iron.			
5	A	124, 282, 301	.....	Cold blast charcoal iron	American as cast	A	.....	Coarse flake graphite } in cold blast charcoal pig iron.
"	AA	124, 282,	.....		Swedish as cast	AA	.....	Scanty and compact graphite }
"	B				Heated to 900° for 15 minutes, furnace cooled	B	33	Primary austenite transformed into pearlite meshes within ferrite network U,U'.
"	C	121,	3.16	Hypo-eutectic gray cast iron	Heated to 900° for 1 hour, furnace cooled	C	"	The graphite and ferrite from the graphitization of the eutectic retain the eutectic shape.
"	D	295			Heated to 900° for 1 hour, and air cooled	D	"	Air cooling under-accents graphitization. This piece when later heated to 900° and furnace cooled became almost like E.
"	E				Heated to 900° for 1 hour, and furnace cooled	E	"	Furnace cooling over-accents graphitization.
					Plate 6. Austenite, cleavage massing, and graphite.			
6	A	187	2.954	Gray cast iron	Solidified very slowly. Heated five times to about 800°, furnace cooled, held at 635° to 750° for 120 hours, and furnace cooled.	A	26	White primaustenoid dendrites in dark eutectic matrix, sodium picrate. Same specimen as AUTHOR IV, Fig. 9, p. 468; *Howe and Levy* IV, Figs. C and D, Row 14; *Author* V, p. 1118. See note 5, p. 603.
"	B	"	3.5	Hypo-eutectic white cast iron	As cast	B	31	Black primaustenoid dendrites in zebra eutectic matrix. *W. Campbell. Author* I, J of Fig. 66.
"	C	"	3.45		Poured into water while molten	C	.....	Ditto. HADFIELD AND HOPKINSON I, Fig. 17b, p. 260.
"	D	184,734	.....	Medium-carbon steel casting.	A solidification fracture	D	42	Primary austenite dendrites.
"	E	106	.....	Rock eutectic		E	.....	Eutectic of quartz, and orthoclase, surrounding primary crystals of acidic plagioclase, in a pre-cambrian trachyte, from near St. Johns, N. B. *Kemp.*
"	F	101	3.16	Coarse Gray cast-	As cast	F	33	Interior of gun-metal pyramid. Solidification structure masked by graphitization.
"	G	101,302	.....	Fine iron casting	" "	G	33	Crust of gun-metal pyramid. Solidification structure indicated by some eutectic.
					Plate 7. Graphite (Figs. C, D, F, G) and undue eutectic (Figs. A, B, E).			
7	A	209	1.08±	Slightly-hyper-eutectic steel ingot	As cast	A	.....	Undue eutectic from segregation on rapid solidification } Small magnification. *Author* III, Fig. 3, p. 645.
"	B	"	"		" "	B	.....	} Large magnification, same specimen as Fig. A. *Author* III, Fig. 5, p. 645.
"	C	118	2.30±	Black-heart malleable castings.	As initially annealed	C	.....	White = ferrite, black = conglomerate of graphite + ferrite. TOUCEDA.
"	D	289	1.45	Very high carbon steel.	Heated to 1,100° for 1 hour, furnace cooled, heated to 1,000° for 1 hour, furnace cooled.	D	24	White ferrite rings about black graphite centers.
"	E	209	.....	High-carbon steel	Solidified rapidly against metal plates	E	29	Undue eutectic from segregation on rapid solidification.
"	F	280	.....	Gray cast iron	As cast	F	32	Moldenke, good gun metal. Black = graphite, white = eutectic, gray = pearlite.
"	G	108, 289	1.45	Very high carbon steel	Heated to 1,100° for 1 hour, furnace cooled, heated to 1,000° for 1 hour, furnace cooled.	G	24	White ferrite borders and streaks about black graphitic areas. *Howe and Levy* V, Fig. 16, Plate 2, and p. 594.
					Plate 8. Progress of the transformation shown micrographically. Hypo-eutectoid and eutectoid steels.			
8	A to E	233	0.40	Medium-carbon steel	Heated to 1,000°, furnace cooled to temperature indicated, and quenched	A to E	9	Progress of the precipitation and coalescence of pro-eutectoid ferrite in extremely slow cooling through the transformation range. *Author* V, Figs. 18 to 22, Plate 1. *Author* VI, Figs. 9 to 11, p. 22. See note 6, p. 603.
"	F to H	223	0.92	Eutectoid steel	Heated to 1,200°±, furnace cooled, reheated differentially and quenched	F to H	14	Transformation of euctectoid steel in rising past the transformation range. See note 7, p. 603.
"	I to N	233	0.40	The same medium carbon steel	Coarsened, then differentially cooled and quenched	I to N	9	Progress of the precipitation and coalescence of ferrite in cooling through the transformation range. *Author* V, Figs. 24 to 28, Plate 1, p. 295. See note 8, p. 604.
"	O to Q	226, 233	0.40		Heated to 1,000°, furnace cooled to temperature indicated, held 30 minutes and quenched.	O to Q	9	Early stages of the precipitation of ferrite in cooling slowly into the transformation range. *Howe and Levy* IV, Figs. A, B, C of Row 11, Plate 4, and pp. 591 and 596. See note 9, p. 604.
"	R to Y	234	0.40		Heated to 1,350°, furnace cooled, reheated differentially, and quenched	R to Y	9	Progress of the reabsorption of ferrite in rising slowly through the transformation range. *Howe and Levy* IV, Figs. A to H of Row 13, Plate 5, and p. 601. See note 10, p. 604.
					Plate 9. Progress of the transformation shown micrographically. Hyper- and hypo-eutectoid steels.			
9	A to H	262	1.45	The same very high carbon steel	Heated to 1,200°, furnace cooled to temperature indicated, and quenched	A to H	24	Progress of the precipitation of pro-eutectoid cementite in cooling slowly through the transformation range. *Howe and Levy* II, Figs. A to F, Plate 1. See note 11, p. 605.
"	I to N	262	1.45		Heated to 1,200°, furnace cooled, reheated except I, to temperature indicated, and quenched.	I to N	24	Progress of the reabsorption of pro-eutectoid cementite in rising through the transformation range. See note 12, p. 606.
"	P to T	174, 257	0.214	Low-carbon steel	Heated to 1,000° for 45 minutes, furnace cooled to temperature indicated, and quenched.	P to T	4	Progress of the precipitation of pro-eutectoid ferrite in a low-carbon steel in cooling slowly through the transformation range. See note 13, p. 606.
					Plate 10. Incipient fusion in steel of 1.60 carbon. Zimmerschied.			
10	A	264	1.6	Very high carbon steel	Heated slightly above solidus and quenched	A	.....	Incipient fusion at austenite grain corners. *Zimmerschied* I, Fig. 15, p. 651.
"	B	265	1.6		Heated further above solidus and quenched	B	.....	Eutectic due to defective diffusion. *Zimmerschied* I, Fig. 17, p. 653.

TABLE 38.—DESCRIPTION OF THE SPECIMENS SHOWN IN THE MICROGRAPHS OF PLATES 1 TO 45.—*Continued*

Plate	Fig.	Referred to in §	Carbon, per cent.	Material	Treatment	Fig.	No. in Table 39	Description and reference
Col. 1	2	3	4	5	6	7	8	9
					Plate 11. Progressive precipitation of pro-eutectoid cementite in hypo-eutectic white cast iron in cooling through the transformation range.			
11	A	272	.....	Moderately hypo-eutectic white cast iron same as Fig. K of Plate 3	Heated to 1,050° for 45 minutes, and quenched	A	28	No precipitation.
"	B	"	.....		" " " " " " furnace cooled to 765°, and quenched	B	28	Precipitation begun; parallel white needles of pro-eutectoid cementite.
"	C	"	.....		" " " " " " " " to 728°, " "	C	28	" continues; more needles; matrix troostite.
"	D	"	.....		" " " " " " " " to 707°, " "	D	28	" " " " ; matrix divorced pearlite.
"	E	"	.....		" " " " " " " " to 671°, " "	E	28	No additional precipitation.
"	F	"	.....		" " " " " " " " to 646°, " "	F	28	" " "
					Plate 12. Cast iron, miscellaneous.			
12	A	269	.....	Spiegeleisen	Heated to 1,050° for 1 hour, and quenched	A	40	Eutectic before transformation. White = eutectic cementite; dark = eutectic austenite.
"	B	"	.....	"	As cast	B	40	" after " White = eutectic cementite; dark = pearlite from eutectic austenite. Same piece as Figs. F and G, Plate 1.
"	C	272	.....	Hypo-eutectic white cast iron, washed metal.	Heated to 1,050° for 1 hour, furnace cooled	C	31	Pro-eutectoid cementite in hypo-eutectic white cast iron. G, white islands = eutectic cementite, large male areas. H = primaustenoid crossed by parallel pro-eutectoid cementite needles thickened by pearlite divorce. Same metal as Fig. B, Plate 6.
"	D	274	3.978	Hypo-eutectic cast iron	As cast. Si, 0.89 per cent.	D	34	Typical eutectic areas G in dark primaustenoid ground mass.
"	E	289	.....	Fine-grained industrial iron casting	As cooled rather rapidly when cast	E	33	Fine: Graphite flakes in industrial casting: Fine-grained radiator casting.
"	F	125, 289 306	3.10	Coke cast-iron wheel	As cast	F	.....	Coarse: Graphite flakes in industrial casting: "Plate" of chilled cast-iron railroad car wheel. See note 14, p. 606.
					Plate 13. Graphitization increases with the slowness of cooling. Chilled cast-iron car wheel.			
13	A	304	.....	Chilled face of tread	As cast	A	37	Dark primaustenoid dendrites in matrix of white eutectic and eutectoid cementite *G*. Small magnification.
"	B	304	.....	" "	" "	B	37	" " Large magnification.
"	C	304, 123	.....	Mottled part about ¾ in. back from face	" "	C	37	Central graphitic region *H*, outer ungraphitized region like Fig. A. Small magnification.
"	D	304, 123	.....	" " " "	" "	D	37	Central area *H* of Fig. C, black graphite streaks in pearlite matrix, imperfect graphitization. Large magnification.
"	E	304	.....	Gray interior of wheel	" "	E	38	Black = graphite flakes; light = pearlite ground mass; white = eutectic islands *T*. Small magnification.
"	F	274, 304	.....	" "	" "	F	38	Lamellar pearlite *S*, honeycombed eutectic island *T*, gray = slag areas *V*. Large magnification.
					Plate 14. Cast iron and miscellaneous.			
14	A	274	3.055	Hypo-eutectic white cast iron.	Solidified quickly by pouring on cold metal plates	A	.....	Borders of undercooling cementite separate primary austenite from the eutectic. *Howe and Levy* VI, Plate 7.
"	B	333, 704	.....	Liquid crystals	Exposed to varying temperature or draughts	B	.....	*Lehmann* I, Fig. 169 on p. 279.
"	C	123, 125	.....	Bad coke cast iron for chilled car wheels.	None	C, D, E	.....	From a single field; illustrate a ferrite skeleton within a graphite flake.
"	D							
"	E							
"	F	50, 766	.....	Wrought iron	Cold rolled 50 per cent. reduction, reheated to 800° for 1 hour, and air cooled	F	.....	Shows the polyhedral structure and the slag banding.
					Plate 15. Idiomorphic and dendritic crystals and cubic cleavages in iron.			
15	A	342	.....	Iron reduced from $FeCl_2$ vapor	Reduced by hydrogen at 685° ±	A	.....	Cubes of ferrite, OSMOND AND CARTAUD I, Fig. 1, Plate II, and p. 120. *Idem* II, Fig. 2, p. 125.
"	B	342	.....	" "	Reduced by zinc vapor at above 1,000°	B	.....	Dendritic cubes of austenite. *Osmond and Cartaud* I, Fig. 18, Plate III, and p. 128. *Idem* II, Fig. 31, p. 136.
"	C	343, 762, 796	.....	Car wheel (rolled)		C	.....	ANONYMOUS
"	D	342	.....	Pig iron	Slowly cooled	D	.....	Cementite plates. *Stead* II, Fig. 25, on p. 176 and p. 182.
"	E	343	.....	Pitchstone, Arran		E	.....	Dendritic crystal of augite in pitchstone. *Rosenbusch* I, Fig. 3 of Plate 3.
"	F	738	0.05	Silicon steel bar	Strained cold, annealed in H above 1,200°, by Ruder	F	.....	Cubic fracture in silicon steel.
"	G	365, 449, 738	Very low	Thermit iron	Solidified slowly from a temperature far above the liquidus	G	.....	" " " ferrite.
					Plate 16. Crystalline structure (A, B, C, E) and expulsion of pro-eutectoid cementite (D, F).			
16	A	345	0.05	Silicon steel sheet	Like Fig. F of Plate 15	A	70	Extremely coarse structure.
"	B	345, 361, 371	0.97	Austenite	Etched at 1,050°	B	.....	*Wark* I, Fig. 727, after p. 730.
"	C	344	.....	Antimony	As cast	C	.....	Dendrites on the surface of an antimony ingot. CAMPBELL.
"	D	368	1.24	Manganese steel	Quenched from 1,100°, polished, and strained cold, reheated to 575° for 2 hours in H, and furnace cooled. Not repolished.	D	60	Expulsion of Fe_3C masks pre-existing slip bands.
"	E	371	0.46	Hypo-eutectoid steel	Heated to 1,200° for 30 minutes, furnace cooled to 730°, and quenched in water	E	9.1	Same specimen as B of Plate 2 but more deeply etched.
"	F	368, 405	1.24	Manganese steel	Like Fig. D of Plate 16, but repolished in the cold	F	60	Near edge. *Howe and Levy* VII, Fig. 5.

TABLE 38.—DESCRIPTION OF THE SPECIMENS SHOWN IN THE MICROGRAPHS OF PLATES 1 TO 45.—*Continued*

Plate	Fig.	Referred to in §	Carbon, per cent.	Material	Treatment	Fig.	No. in Table 39	Description and reference
Col. 1	2	3	4	5	6	7	8	9
					Plate 17. Expulsion of the pro-eutectoid element (A, B, C, D, F). Etching figures in ferrite (E, G, H).			
17	A	368	1.24	Hadfield's manganese steel.	Quenched from 1,100°, held 2 hours at 575° in H. Copper plated and cross-sectioned. Polished cold.	A	60	Surface coated with pro-eutectoid cementite.
"	B					B	"	Interior cavities coated with cementite.
"	C	367	.....	Vug crystal	As cast	C	.....	Slag lines in ferrite.
"	D	368	1.45	Hyper-eutectoid steel	Heated to 1,000° in sealed quartz tube, furnace cooled, and copper plated	D	24	Surface coated with pro-eutectoid cementite. Fe_3C sheath at edge.
"	E	371, 374	Trace.	Embrittled dead soft steel.	Deeply etched	E	.....	Dyke-like protrusions of ferrite from ferrite. *Stead* II, Fig. 19, p. 176.
"	F	368	1.45	Hyper-eutectoid steel	Same specimen as Fig. D of this plate	F	24	Surface coated with pro-eutectoid cementite.
"	G	371, 374	.....	Ferrite, silicon steel	Si 4.5	G	.....	Deep etching discloses cubic cleavages in ferrite. *Stead* II, Fig. 18.
"	H	371	0.02	" " "	Si 4.0	H	.....	Cubic etching pits in silicon steel. *Stead* II, Fig. 15.
					Plate 18. Etching and pressure pits in ferrite.			
18	A	108, 373, 532	0.40	Hypo-eutectoid steel	Heated to 1,000° ±, furnace cooled. Punched	A	9	Ferrite and pearlite hairs caused by punching. *Howe and Levy* V, Fig. 8.
"	B	108, 371, 373	0.40	" " "	" " " " " " "	B	"	Details of Fig. A. Geometrical etching pits after extreme deformation.
"	C	108, 371, 373	0.40	" " "	" " " " " " "	C	"	
"	D	108, 371	0.01	A.R.M. steel	Heated to 1,300° ±, and air cooled	D	0	Shape of etching pits changes from grain to grain.
"	E	371, 373	.....	Swedish wrought iron	Cold rolled from ¾ in. to ½ in. diameter	E	.....	Cubic etching pits in cold rolled iron. *Ewing and Rosenhain* I, Fig. 33, Ewing I, Fig. 4, Plate 3.
"	F	379	.....	Iron, ferrite	Cube face punctured and pressed on a smooth surface	F	.....	Deformation converts initial round holes into squares. Stead.
"	G	371	.....	" "	Cube face pricked with needle and strongly etched	G	.....	Etching squares a rounded hole. Stead.
"	H	776	.....	.....	Large ingot (about 400 to 500 lb.)	H	.....	Abrupt change from columnar to equiaxed grains. Columnar structure.
					Plate 19. Lüders lines. Hartmann.			
19	A	391	.....	Steel	Polished and dished	A	.....	Pattern of puckerings. *Hartmann* II, Fig. 31.
"	B	"	.....	"	Polished and strained tensilely	B	.....	Lines of direct system circumferential and transverse to stress, and lines of oblique system. *Hartmann* I, Plate 5.
"	C	"	.....	"	Key to Fig. B	C	.....	*Hartmann* I, Fig. 142.
					Plate 20. Lüders lines.			
20	A	391	.....	Glass	Perforated by bullet	A	.....	Radial and circular lines of rupture.
"	B	396, 399	.....	Very coarse-grained silicon steel sheet	Strained cold, annealed in hydrogen above 1,200° by Ruder. (Like F of Plate 15) — Broken tensilely	B	.....	Sawteeth near B of Fig. D.
"	C	399	.....		— Strained tensilely	C	.....	Beginning of path of rupture.
"	D	396, 399	.....		— Broken tensilely	D	.....	Rupture follows conjugate planes.
"	E	395	.....	Wood	Squeezed in vise	E	.....	Lüders lines in wood.
"	F	397	.....	Thermit iron	Wedge squeezed in vise	F	0.5	Wedge badly deformed.
					Plate 21. Grain uplift in ferrite and austenite.			
21	A	410	0.05	Silicon steel	Like Figs. B to D of Plate 20. Polished, squeezed in vise	A	66	Accordion markings.
"	AA	408	1.24	Hadfield's manganese steel casting	Heated to 1,100° quenched, polished, hammered	AA	"	Persistence of columnar structure through 1,100° quenching.
"	B	411			Casting not reheated, polished, hammered	B	"	Surface crumplings in manganese steel. Slip bands change direction at boundaries of columnar grains in manganese steel.
"	C	405, 408			Casting not reheated, polished, hammered	C	"	
"	D	407			Heated to 1,300° ±, air cooled, polished, squeezed, copper plated	D	0	Cross-section through a copper-plated specimen showing grain uplift.
"	E	407	0.01	A.R.M. steel	As received. Polished, squeezed	E	0	Grain uplift on previously polished surfaces.
"	F	407			Like D	F	0	Grain uplift.
"	G	407, 408			Like E	G	0	
"	H	417	0.2	Invar (Ni 36 per cent.)	Rolled bar, polished, squeezed	H	65	Four directions of slip bands. Invar.
					Plate 22. Slip bands in ferrite and miscellaneous.			
22	A	380	0.01	A.R.M. steel	As received	A	0	Same spot — Vertical light.
"	B	380	0.01	" " "	" "	B	0	Same spot — Oblique light.
"	C	412	Very small.	Thermit iron	" " Polished, squeezed in vise	C	0.5	Irregular slip bands in ferrite.
"	D	412	0.01	A.R.M. steel	" " " " " "	D	0	Slip bands retain their straightness close to grain boundaries.
"	E	412, 498	.....	Thermit iron	Like C	E	0.5	Very irregular slip bands.
"	F	408, 412	0.01	A.R.M. steel	Heated to 1,300° ±, air cooled, polished, squeezed in vise	F	0	Slip bands accent the position of grain boundaries.
"	G	410	1.24	Manganese steel	" " 1,100° for 2 hours and quenched. Polished and bent	G	60	Sausage-like projections in austenite.

TABLE 38.—DESCRIPTION OF THE SPECIMENS SHOWN IN THE MICROGRAPHS OF PLATES 1 TO 45.—*Continued*

Plate	Fig.	Referred to in §	Carbon, per cent.	Material	Treatment	Fig.	No. in Table 39	Description and reference
Col. 1	2	3	4	5	6	7	8	9
Plate 23. Slip bands in ferrite (all etched except B).								
23	A	412, 497	0.03	O. H. steel	As received, polished: etched, squeezed in vise	A	1	Slip bands in etched ferrite.
"	B	412,			not etched, squeezed in vise	B	"	" " fault each other.
"	C	412,			etched, squeezed in vise	C	"	" " in etched ferrite.
"	D	412, 497			" " " "	D	"	" " cross grain boundaries in etched ferrite.
"	E	412, 503			" " " "	E	"	" " curl to cross grain boundaries.
"	F	412,			" " " "	F	"	" " not reaching terminal boundary.
Plate 24. Slip bands and grain uplift in ferrite, austenite, and copper.								
24	A	498	.....	Electrolytic plate of copper	Polished and squeezed in vise	A	.....	Slip bands in twinned areas in copper.
"	B	425	0.01	A.R.M. steel	Heated to 1,300° ±, air cooled, polished, etched, squeezed in vise	B	0	Slip bands crossing grain fragments in etched ferrite.
"	C	408	0.01	" " " "	" " " " , air cooled, polished, not etched, squeezed in vise	C	"	Slip bands indicate the grain boundaries not otherwise indicated.
"	D	408	1.24	Manganese steel	" " 1,100°, quenched, polished wedge, not etched, squeezed in vise	D	60	Grain uplift and slip bands in austenite.
"	E	408, 412	0.01	A.R.M. steel	Like B	E	0	Unexplained geometrical arrangement of uplift.
"	F	408, 412	0.01	" " " "	Like C	F	"	Slip bands cross uplift grain boundaries.
Plate 25. Slip bands from repetitive stress reversals, and cross-section through slip bands.								
25	A	422	.....	Swedish wrought iron	Strained repeatedly	A	.....	5,000 reversals, a few slip bands.
"	B	422, 497				B	.....	40,000 " many slip bands, some broad black areas
"	C					C	.....	60,000 " resembles B.
"	D	422				D	.....	70,000 " repolished and etched.
"	E	418, 466	Very small.	Sheet steel	Polished, strained in tension, copper plated	E	.....	Cross-section through slip bands. *Rosenhain* I, Fig. 5, after p. 224.
"	F	407, 418, 466	0.01	A.R.M. steel	Heated to 1,300° ±, air cooled, polished, squeezed in vise, copper plated	F	0	Cross-sections through polished, deformed, and copper-plated ferrite.
"	G	418, 466				G	"	
"	H	407, 466				H	"	
"	I	418, 466				I	"	

Ewing and Humfrey I, Figs. 3 to 6.

Plate	Fig.	Referred to in §	Carbon, per cent.	Material	Treatment	Fig.	No. in Table 39	Description and reference
Plate 26. Deformation in carbon steel and wrought iron, and silhouette.								
26	A	108, 538, 680	0.214	Low-carbon steel	Heated to 1,100° for 1 hour, furnace cooled; heated to 900° for 1 hour, furnace cooled, and pulled in tensile test. Like Fig. F of Plate 2.	A	4	Rupture of divorced cementite in steel of 0.21 per cent. C.
"	B	532, 537, 680	2.04	Hyper-eutectoid steel	Bent at dull red heat till it began to crack.	B	25.1	Rupture of cementite masses in steel of 2.04 per cent. C.
"	C	108, 766	Small.	Muck bar	Bent double at white heat	C	.....	Flow and rupture of slag in bending wrought iron while hot.
"	D	680, 721, 766	Small.	Wrought iron	Punched by Hoopes and Townsend	D	.....	Rupture of slag and inflow of ferrite in punching wrought iron.
"	E	451	0.05 or 0.06	Crystalline iron	Needle pricked, then bent gently. Cube face	E	.....	Irrelation between silhouette in ferrite and later slip bands. After *Osmond and Cartaud* III, Fig. 15, p. 461.
"	F	541, 680	0.40	Medium-carbon steel	Deformation in punching after furnace cooling from 1,000° ±	F	9	Movement of pearlitic cementite sheets in punching. *Howe and Levy* V, Fig. 21.
Plate 27. Deformation on previously polished surfaces of carbon steel. (See Note 15, p. 606.)								
27	A	532	0.92	Eutectoid steel	Heated to 900°, furnace cooled to 680°, air cooled. Polished, squeezed in vise	A	14	Extrusion of pearlitic ferrite by pressure. C 0.92.
"	B	531, 543	0.40	Medium-carbon steel	Heated 2 hours ± to 1,200°, air cooled	B	9	Extrusion of ferrite network GE and EF by pressure. C 0.40.
"	C	543			Polished, squeezed in vise	C	"	Deformation lines in sorbite simulating slip bands.
"	D	543			" " " "	D	"	
"	E	531, 725	0.59		Heated 3 hours to 1,300°, air cooled. Polished, squeezed in vise	E	11	Extrusion of ferrite network by pressure. Oblique light. C 0.59.
"	F	531, 543, 725				F	"	Extrusion of ferrite network *ab* and pearlitic ferrite in steel of 0.59 C by pressure. Same as E.
Plate 28. Deformation in carbon steel, and Widmanstättian figuring.								
28	A	235, 531, 718	0.40	Medium-carbon steel	Polished, etched	A	9	Widmanstättian figuring. Same as B of Plate 27.
"	B	531, 718,			Heated 2 hours ± to 1,200°, furnace cooled: Polished, etched	B	"	Atolls of undercooling cementite about ferrite beams of Fig. A.
"	C	531			Polished, not etched, squeezed in vise.	C	"	Extrusion of Widmanstättian ferrite beams of Fig. A by pressure. Unetched. C 0.40.
"	D	531, 543			Polished, not etched, squeezed in vise.	D	"	Slip bands in the ferrite zones of Fig. A.
"	E	545			Heated 15 minutes to 1,100° and quenched. Polished, not etched, squeezed in vise.	E	"	Extrusion of martensitic markings by pressure in hardened steel. Black zone in E is a crack.
"	F	545				F	"	

TABLE 38.—DESCRIPTION OF THE SPECIMENS SHOWN IN THE MICROGRAPHS OF PLATES 1 TO 45.—*Continued*

Plate	Fig.	Referred to in §	Carbon, per cent.	Material	Treatment	Fig.	No. in Table 39	Description and reference
Col. 1	2	3	4	5	6	7	8	9
Plate 29. Annealing twins in austenite.								
29	A	558	1.24	Manganese steel	Tensile fracture, heated 1 hour to 1,000°, and quenched. Polished, not etched, hammered.	A	60	Reversal of relative prominence of conjugate sets of slip bands in twinned zones.
"	B	412				B	"	Zigzagging of a single set of slip bands in crossing twinned zones.
"	C	558, 560, 566			Heated to 1,100° and quenched, polished, not etched, strained in vise	C	"	" " " " " " " " " a very narrow twinned zone.
"	D	412	0.2	Invar	Heated 30 minutes to 1,200°, air cooled, squeezed strongly in vise, heated 15 minutes to 600°, air cooled, polished, squeezed gently in vise.	D	65	Almost complete effacement of slip bands in crossing a twinned zone.
"	E	558, 560				E	"	Reversal of relative prominence of conjugate sets of slip bands in twinned zones.
"	F	558				F	"	Like B, except that strength of contrast between treads and risers changes little.
Plate 30. Twins in ferrite, austenite, and minerals (§§546 to 569).								
30	A	547	.....	Carrara marble	None	A	.....	Mechanical twins. *Adams and Nicolson* I, Fig. 1 of Plate 24, after p. 401.
"	AA	501, 558	.....	Native copper crystal	Polished and very slightly deformed	AA	.....	Twin in native copper crystal very slightly deformed.
"	AB	" "	.....	" " "	" " deformed more	AB	.....	The same twin after more deformation.
"	B	552	.....	Diabase		B	.....	Twin striation of plagioclase according to albite law, *Rosenbusch* I, Fig. 6 of Plate 25.
"	C	546, 561	0.2	Invar	Like Figs. D to F of Plate 29	C	65	A single field in two positions under oblique light. Changing the incidence of light reverses the relative brightness of the twinned zones.
"	D	" "				D	65	
"	E	639	.....	Thermit 33 per cent. MnFe alloy.	Heated to 1,100°, quenched, heated 2 hours to 550°, air cooled, squeezed gently in vise.	E	61	Twinning does not deflect grain boundary.
"	F	558, 600	Small.	Thermit ferrite	As received, polished, not etched, squeezed in vise	F	0.5	Probable mechanical twins in ferrite.
Plate 31. Slip bands and Neumann bands (§§570 to 609).								
31	A	410, 571	0.01	A.R.M. steel	Heated to 1,300° ±, air cooled. Polished, not etched, hammered	A	0	Curves in probable Neumann bands.
"	B	" "				B	"	Slip band projections in ferrite.
"	C	571			" etched, hammered	C	"	Neumann bands on a previously etched surface deflect at grain boundaries.
"	D	503, 588				D	"	" and slip bands.
"	E	571			" not etched, hammered	E	"	" bands under oblique light.
"	F	"				F	"	" bands (?).
Plate 32. Neumann bands on unetched surfaces of ferrite (silicon steel) (§§570 to 609).								
32	A	409, 571	0.05	Ruder's crystalline silicon steel.	As received, *i.e.*, strained cold, annealed in H above 1,300°. Polished, not etched, squeezed in vise.	A	66	Grain uplift (?) and Neumann bands.
"	B	" "				B	"	Curvature caused by faulting.
"	C	571				C	"	Slight faulting of intersecting Neumann bands.
"	D	"			Polished, not etched, hammered	D	"	Rupture follows the directions of two sets of Neumann bands.
"	E	409, 571				E	"	Curved Neumann bands.
Plate 33. Neumann bands developed on polished surfaces by etching (§§570 to 609).								
33	A	600, 571	0.01	A.R.M. steel	Hammered, polished, etched, squeezed in vise	A	0	Probably accidental zigzagging of slip bands on crossing a Neumann band.
"	B	571	0.05	Silicon steel	As received, *i.e.*, like Plate 32. Hammered, polished, etched, squeezed in vise	B	66	Ladder pattern.
"	C	576	0.01	A.R.M. steel	Like A	C	0	Faulting of Neumann by slip bands.
"	D	588, 605	0.05	Silicon steel	" B	D	66	Neumann *MM'* faults *NN'*.
"	E	573	0.01	A.R.M. steel	" A	E	0	Slip bands cross Neumann bands (?).
"	F	424, 588	0.05	Silicon steel	" B	F	66	Discontinuous slip bands (?).
Plate 34. Neumann bands developed on polished surfaces by etching.								
34	A	590	0.05	Crystalline silicon steel.	As received, *i.e.*, like Plate 32. Cube face of fracture, not polished, needle pricked in one place, etched deeply.	A	66	Nine directions for Neumann bands on cube face.
"	B	604	0.01	A.R.M. steel	Heated to 1,300° ±, air cooled, hammered, polished, etched	B	0	Varying degrees of nodality in etched Neumann bands.
"	C	590	.....	.....	Key to A. Cube face is turned 45° from usual position	C	.....	Approximation of Neumann bands to theoretical directions.
"	D	605	0.05	Crystalline silicon steel.	As received, *i.e.*, like Plate 32. Hammered, polished, etched, squeezed in vise	D	66	Crystallographic serrations in a Neumann band developed by etching.
"	E	571	0.01	A.R.M. steel	Like B, but etched with Heyn's reagent	E	0	Spotty widening of Neumann bands developed by etching.
"	F	580	0.05	Crystalline silicon steel.	Detail of A	F	66	Six directions of Neumann bands on cube face. Details of Fig. A.
Plate 35. Neumann bands on pre-etched surfaces (A), on cubic cleavages (B, C, D), and deepened by etching on cubic cleavages (E, F, G).								
35	A	604	.....	Cast zinc	Bent after casting	A	.....	Neumann bands resume initial direction on re-entering a given grain. *Campbell* I, Fig. 931, after p. 808.
"	B	587, 590	Low.	Thermit iron, A of Fig. G of Plate 15.	Cubic cleavages not polished or etched. Vertical light	B	0.5	Acanthus figuring. Cubic and Neumann bands on cube face.
"	C				Oblique light	C	0.5	Neumann bands follow the 211 directions on unetched cubic fracture faces.
"	D				Oblique light	D	0.5	
"	E	371	Low.	Thermit iron, another fragment.	Deeply etched without polishing	E	0.5	Deeply etched Neumann bands follow the 211 directions on three adjacent faces of one prism.
"	F	371, 580				F	0.5	
"	G	" "				G	0.5	

TABLE 38.—DESCRIPTION OF THE SPECIMENS SHOWN IN THE MICROGRAPHS OF PLATES 1 TO 45.—*Continued*

Plate	Fig.	Referred to in §	Carbon, per cent.	Material	Treatment	Fig.	No. in Table 39	Description and reference
Col. 1	2	3	4	5	6	7	8	9
					Plate 36. Neumann bands and X bands.			
36	A	466, 602, 612	0.05	Ruder's crystalline silicon steel as received (Plate 32)	Polished, not etched, squeezed in vise, copper plated, cross-sectioned. (Same piece as A,B,C of Plate 32).	A	66	Profiles of Neumann bands on a cross-section after copper plating.
"	B					B	"	
"	C					C	"	
"	D					D	"	
"	E	580, 590	.....	Thermit iron	Same piece as E, F, G, of Plate 35. Copper plated, cross-sectioned parallel to one of the diagonals of E, i.e., along a 011 plane.	E	0.5	Profile of etching furrows of E, Plate 35.
"	F					F	0.5	Neumanns follow 211 directions on rhombo-dodecahedral (011) faces.
"	G	623	0.01	A.R.M. steel.....	Heated to 1,300° ±, air cooled, punched ⅝–¾ in. out of 1-in. plate, polished and deeply etched.	G	0	X bands developed by repolishing and etching deformed ferrite.
"	H	574	.....	Thermit iron	Detail of E and F, this plate	H	0.5	The nodes of etched Neumann bands avoid each other.
"	I					I	0.5	Parallel banding of etched Neumann lamellæ.
					Plate 37. X bands in ferrite.			
37	A	623	0.01	A.R.M. steel.....	No thermal treatment, compression 30,000 lb. per sq. in. — etched after 2 hr. rest.....	A	0	Strong X bands in certain grains.
"	B				" " 49 hr. rest.....	B	"	
"	C				" " 38 days rest.....	C	"	
"	D				" " 24 hr. rest+15 min. at 100°	D	"	
"	E				" " 24 hr. " +15 " " 300°	E	"	
"	F	623, 635	0.01	A.R.M. steel.....	Punching, same as Fig. G of Plate 36 — " " 15 min. at 200°.....	F	"	
"	G				" " 15 " " 600°.....	G	"	
					Plate 38. Deformation lines in manganese steel and copper.			
38	A	640, 646, 653, 655	1.24	Hadfield's manganese steel casting	Quenched from 1,100°, strained slightly by bending, heated 2 hours to 575°, etched lightly.	A	60	Pro-eutectoid cementite precipitated in octahedral cleavages and grain boundaries on heating. *Howe and Levy* VII, Fig. 13
"	B	662			Quenched from 1,100°. Wedge, polished, not etched, squeezed in vise.....	B	"	Strongly marked slip bands. *Idem*, Fig. 16.
"	C	412, 662	.....	Lake Superior copper, rolled bar	Wedge polished, squeezed in vise, not etched.....	C	.....	Strongly marked slip bands. *Idem*, Fig. 20.
"	D	653, 655, 662	1.24	Hadfield's manganese steel casting	Like A.....	D	"	Cementite balled up in octahedral cleavages, *Idem*, Fig. 14.
"	E	662			Like B.....	E	"	Marked slip bands. *Idem*, Fig. 17.
"	F	412, 662	.....	Copper, like C.....	Like C.....	F	.....	Marked slip bands. *Idem*, Fig. 21.
"	G	656	1.24	Hadfield's manganese steel casting	Quenched from 1,100°, etched deeply, not cleaned.....	G	"	Eutectiform pattern in the interdendritic fillings.
"	H	662			Like B.....	H	"	Disturbance of two sets of slip bands at their intersections. *Idem*, Fig. 18.
"	I	412, 662	.....	Copper, like C.....	Like C.....	I	.....	Two sets of slip bands in copper. *Idem*, Fig. 22.
"	K	662	1.24	Hadfield's manganese steel casting	Like B.....	K	"	Incipient slip bands in manganese steel. *Idem*, Fig. 19.
"	L	412, 662	.....	Copper, like C.....	Like C.....	L	.....	Incipient slip bands in copper. *Idem*, Fig. 23.
					Plate 39. Deformation lines in manganese steel.			
39	A	640, 645	1.24	Hadfield's manganese steel castings, quenched from 1,100°	Deeply etched, then cleaned.....	A	60	Curly etching lines like slip bands. *Howe and Levy* VII, Fig. 3.
"	B	657			Like G of Plate 38. Deeply etched — then cleaned.....	B	"	Straight octahedral lines after etching.
"	C	657			Like G of Plate 38. Deeply etched — then partly cleaned.....	C	"	Deformation of the eutectiform pattern by cleaning.
"	D	660			Polished, lightly strained, not etched.....	D	"	Conjugate slip bands reversing relative prominence in certain zones. *Idem*, Fig. 6.
"	E	412, 641			Tensile fracture heated 1 hour to 1,000°, quenched, polished, not etched, hammered. Like A and B of Plate 29.	E	"	Slip bands stop or change prominence at boundaries. *Idem*, Fig. 7.
"	F	656			Deeply etched, not cleaned. Corner of specimen G of Plate 38.....	F	"	Columnar eutectiform etching pattern.
"	G	641			Polished, not etched, hammered.....	G	"	Slip bands cross dendrites and grain boundaries. *Idem*, Fig. 8.
"	H	425, 561, 656			Deeply etched, not cleaned — Part of specimen A of this Plate.....	H	"	Eutectiform pattern.
"	I	656			Deeply etched, not cleaned — Part of specimen F of this Plate.....	I	"	Crackly eutectiform pattern.
					Plate 40. Fracture and path of rupture.			
40	A	692	0.05 ±	Granular silicon steel...	Heated 8 hours to 1,400° in hydrogen by Ruder.....	A	.....	Curved intergranular (chondritic) fracture of silicon steel.
"	B	731	.....	Müsen spiegeleisen.....	As cast.....	B	.....	Cementite plates in fracture of spiegeleisen.
"	C	762	.....	Wrought iron.....	Nicked on one side and bent away from the nick.....	C	.....	Barked wrought iron.
"	D	694	.....	Bismuth.....	As cast, polished, etched, cracked.....	D	.....	Rupture seeks grain centers in bismuth.
"	E	738	.....	Silicon steel.....	Deeply etched.....	E	.....	Rectangular etching figures. Rupture follows cubic cleavages. STEAD.
"	F	740	Small	Soft steel sheet.....	Embrittled.....	F	.....	Rectangular brittleness accompanied by diagonal ductility. *Stead* II, Fig. 24.

TABLE 38.—DESCRIPTION OF THE SPECIMENS SHOWN IN THE MICROGRAPHS OF PLATES 1 TO 45.—*Continued*

Plate	Fig.	Referred to in §	Carbon, per cent.	Material	Treatment	Fig.	No. in Table 39	Description and reference
Col. 1	2	3	4	5	6	7	8	9
					Plate 41. Path of rupture in steel, wrought iron, and bismuth.			
41	A	724	0.01	A.R.M. steel	As rec'd. Tensile fracture copper plated. Longitudinal section, deeply etched	A	0	Drawing out of ferrite prongs in tensile rupture.
"	B	764	Small.	Muck bar	As received. Polished, etched, cracked	B	.....	Rupture follows slag in wrought iron.
"	C	717, 725	0.59	Hypo-eutectoid steel	Heated 3 hours to 1,300°, furnace cooled. Polished, etched, cracked	C	11	Rupture does not conform to roundabout ferrite network.
"	D	570, 609	.....	Bismuth	Cast, polished, etched, cracked	D	.....	Trans-crystalline rupture, Neumann bands, and broad twin, O, in bismuth.
"	E	694, 717, 731	1.45	Hyper-eutectoid steel	Heated to 1,100° ±, furnace cooled. Polished, etched, hammered	E	24	Rupture follows cementite network in hyper-eutectoid steel.
					Plate 42. Path of rupture in steel and nickel. (See Note 16.)			
42	A	717	0.14	Low-carbon steel	Heated 10 minutes to 1,100°, furnace cooled. Polished, etched, cracked	A	17	Rupture chiefly in the ferrite.
"	B	719	0.40	Medium-carbon steel	Heated 2 hours to 1,200°, furnace cooled. Polished, etched, cracked	B	9	In this Widmanstättian structure rupture cannot confine itself to the ferrite.
"	C	689	0.106	Low-carbon steel	Broken at 932° (tensile fracture)	C	.....	Intergranular rupture in yellow-hot austenite. *Rosenhain and Humfrey* I, Fig. 28 and p. 262.
"	D	720	.....	Medium-carbon steel	J. P. Snow's rail cracked in service	D	.....	Rupture passes through pearlite and ferrite nearly indiscriminately.
"	E	731	1.24	Hadfield's manganese steel.	Heated to 1,100°, quenched, reheated 2 hours to 575°, broken	E	60	Great preference of rupture for intergranular pro-eutectoid cementite. *Howe and Levy* VII, Fig. 9.
"	F	717	0.214	Low-carbon steel	Heated 1 hour to 1,300°, furnace cooled. Polished, etched, cracked	F	9	Rupture shows some preference for ferrite.
"	G	762	.....	Thermit nickel	As cast. Broken with hammer	G	.....	Very fibrous fractures in cast nickel. Unforged. Goldschmidt Thermit Co.
"	H	724	0.46	Medium-carbon steel	Heated 10 minutes to 1,100°, fire-clay cooled. Broken, copper plated, etc.	H	9.1	Rupture passes almost indiscriminately across ferrite and pearlite.
"	I	749, 762	0.29	Nickel steel	Heated to 800°, quenched, reheated to 580° ±, air cooled, notched, broken with sledge.	I	.....	Very fibrous fractures in rolled nickel steel.
					Plate 43. Ferrite ghosts.			
					Phosphorus			
43	A	762	0.147	Cold-rolled steel	..... Heated 10 min. to 1,000° → Air cooled	A	17.1	Retarding the cooling increases the prominence of the ghosts.
"	B				..... Longitudinal section → Furnace cooled	B	17.1	
"	C	762	0.34	Phosphoric steel	0.417 Furnace cooled to 650°, then air cooled	C	20	Ferrite ghosts of abnormally phosphoric steel.
"	D	762	0.46	Medium-carbon steel	0.041 Heated 10 min. to 1,100°	D	9.1	Ghosts visible to the naked eye.
"	E	367, 762, 786	0.40		0.014 " " " " 1,100° quenched, reheated → 5 min. to 725°, 30 min. to 690° → air-cooled longitudinal section	E	9	Ferrite banding seen best from a distance.
"	F	762, 780, 786	0.46		0.041 " " " " 800° quenched, reheated → 30 min. to 600° → air-cooled longitudinal section	F	9.1	Banding caused on reheating bandless quenched steel to 600°.
"	G	762	0.40		0.014 " " " " 1,100° quenched, reheated → 10 min. to 725° → air-cooled longitudinal section	G	9	Banding caused on reheating to 725° in almost complete absence of phosphorus.
					Plate 44. Ferrite banding.			
44	A	809			Sheared plate from C. L. Huston → Flat section parallel to face of plate	A	.....	Patterns probably representing inaccuracy of the section.
"	B	"				B	.....	
"	C	"	0.20	Boiler plate steel	Universal plate from C. L. Huston	C	.....	Slight longitudinal banding.
"	D	"				D	.....	
"	E	762			Transverse section of Plate D	E	.....	Slight transverse banding even in a universal plate.
"	F	"			Longitudinal section of Plate B	F	.....	Persistence of longitudinal banding through transverse final passes.
"	G	367, 762, 766	0.43	Medium-carbon steel	Forged at abnormally low temperature	G	.....	Slag rods in ferrite.
"	H	796	0.46	Medium-carbon steel	Heated 10 minutes to 1,100°, quenched, reheated 10 minutes to 1,080°, fire-clay cooled.	H	9.1	Combination of ghosts and network structure.
"	I	679, 765	Small.	Wrought iron	Cold rolled, 50 per cent. reduction, reheated 1 hour to 550°	I	.....	Smoothness of grain outlines even after great deformation.
					Plate 45. Deformation lines in twisted crystals, thorns, and edgings.			
45	A	582, 599	.....	Meteoric iron	Bar cut parallel to crystalline axis, and twisted axially. Longitudinal section polished and etched.	A	.....	Parallel changes in tint in parallel Neumann lamellæ. Stead.
"	B	582	.....			B	.....	Fringes caused by twisting. Stead.
"	C	582	0.08	Extra soft steel		C	.....	Thorns and edgings in ferrite. Divorced cementite in grain boundaries. *Osmond, Fremont, and Cartaud* I, Fig. 31 and pp. 27, 29.
"	D	501, 582, 676	.....	Native copper crystal.	Bar cut parallel to axis of crystal, polished, and gently squeezed. D and F cut after twisting bar 360° about its axis. → longitudinal section	D	.....	Slip bands regular and geometrical in spite of previous deformation.
"	E	412, 417, 441, 501			→ transverse section	E	.....	Unusually regular slip bands.
"	F	582, 676	.....		→ transverse section	F	.....	Twinned zone regular in spite of deformation.

TABLE 39.—COMPOSITION, PROPERTIES, AND SOURCES OF THE STEELS AND IRONS USED IN THE EXPERIMENTS MADE FOR THIS WORK

Designation number	Chemical composition, per cent.							Mechanical properties, after coarsening at 1,100°, grain refining, and annealing 3 hours at 600° (except No. 0, as rolled, and No. 2, annealed 3 hours at 550°)							Source	Analyzed by[a]	Form of specimens
	Carbon			Si	Mn	P	S	Tensile strength, lb. per sq. in. (kg. per sq. mm.)	Yield point, lb. per sq. in. (kg. per sq. mm.)	Proportionality limit, lb. per sq. in. (kg. per sq. mm.)	Elongation in 4 in., per cent.	Contraction of area, per cent.	Fremont test, kg. m. per sq. mm.	Brinell hardness No.			
	Total	Combined	Graphitic														
I. Electrolytic iron, hypo-eutectoid and eutectoid steels, A, normal																	
0	0.01	.....	.....	Very low				44,190 (31.1)	20,180 (14.2)	.....	27.8	68.6	.....	75	Am. Rolling Mill Co.	.....	Billet about 4 in. square.
1	0.023	.....	.....	.....	Nil	.....	.....	.....	.....	.....	.....	.....	.....	.....	C. F. Burgess	B.	Cathode sheet.
2	0.027	.....	.....	0.003	0.26	0.005	0.024	51,040 (35.9)	39,535 (27.8)	38,850 (27.3)	31.5	77.0	0.50 N.B.	92	W. B. Kunhardt	B. & M.	Rolled bar ⅝ in. diam.
2.1	0.05	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	C. S. Marshall & F. H. Daniels	M.	Rolled bar ⅝ in. diam.
3	0.105	.....	.....	0.013	0.24	0.015	0.028	.....	.....	.....	.....	.....	.....	.....	Pittsburg T. L.	P. T. L.	Plate 1 by 5/16 in.
4	0.214	.....	.....	0.039	0.05	0.013	0.010	80,600 (56.7)	61,390 (43.2)	59,900 (42.1)	12.0	68.5	0.36 N.B.	153	W. B. Kunhardt	B. & M.	Rolled bar ⅝ in. diam.
5	0.227	.....	.....	0.039	0.05	0.013	0.010	.....	.....	.....	.....	.....	.....	.....	" " "	B. & M.	Rolled bar ⅝" diam.
6	0.235	.....	.....	0.039	0.05	0.013	0.010	.....	.....	.....	.....	.....	.....	.....	" " "	B. & M.	Rolled bar ⅝" diam.
7	0.244	.....	.....	0.050	Nil	Nil	0.007	.....	.....	.....	.....	.....	.....	.....	C. F. Burgess	B.	Rolled bar ⅝" diam.
8	0.382	.....	.....	0.027	0.22	.....	0.004	.....	.....	.....	.....	.....	.....	.....	W. Campbell	B. & R.G. Morse	Rolled bar ¾ in. diam.
9	0.40	.....	.....	0.103	0.16	0.014	0.012	88,290 (62.1)	65,275 (45.9)	57,800 (40.6)	15.5	57.7	0.203	187	W. B. Kunhardt	B.	Rolled bar ⅝ in. diam.
10	0.563	.....	.....	0.18	0.15	0.013	0.013	.....	.....	.....	.....	.....	.....	.....	J. A. Mathews	B. & M.	Rolled bar ½" diam.
11	0.59	.....	.....	0.144	0.17	0.018	0.013	109,840 (77.2)	74,000 (52.0)	69,650 (49.0)	15.8	49.8	0.056	202	C. S. Marshall & F. H. Daniels	B. & M.	Rolled bar ⅝ in. diam.
12	0.73	.....	.....	0.141	0.07	0.012	0.019	.....	.....	.....	.....	.....	.....	.....	W. Campbell	B. & M.	Rolled bar 5/16" square.
13	0.78	.....	.....	0.108	0.26	0.011	0.013	.....	.....	.....	.....	.....	.....	.....	Pittsburg T. L.	P. T. L.	Plate ⅛ in. thick.
14	0.92	.....	.....	0.14	0.123	0.009	0.011	109,840 (77.2)	74,340 (52.3)	.....	3.8	13.7	0.086	212 223	J. A. Mathews	B. & M.	Plate ⅝ in. diam.
B. High manganese																	
15	0.11	.....	.....	.....	0.67	.....	.....	.....	.....	.....	.....	.....	.....	.....	Dealer	P. T. L.	Rolled bar 1" square.
16	0.110	.....	.....	0.004	0.60	0.036	0.064	.....	.....	.....	.....	.....	.....	.....	W. Campbell	B.	Rolled bar ½" square.
17	0.140	.....	.....	0.040	1.16	0.020	0.018	78,130 (54.9)	60,160 (42.3)	.....	13.8	73.6	0.504 N.B.	126	W. B. Kunhardt	B. & M.	Rolled bar ⅝ in. square.
17.1	0.147	.....	.....	0.018	0.655	.....	.....	.....	.....	.....	.....	.....	.....	.....	Dealer	B.	Cold-drawn ¾" diam.
18	0.32	.....	.....	0.122	0.41	0.004	0.022	.....	.....	.....	.....	.....	.....	.....	Pittsburg T. L.	P. T. L.	Forging ½ in. diam.
19	0.46	.....	.....	0.094	1.215	0.017	0.015	122,720 (86.3)	107,850 (75.8)	101,210 (71.2)	11.8	56.0	0.201	192	W. B. Kunhardt	B.	Rolled bar ⅝ in. diam.
19.1	0.46	.....	.....	.....	0.466	0.041	.....	.....	.....	.....	.....	.....	.....	.....	Anonymous	B.	Rolled bar 1" square.
C. High phosphorus and high sulphur																	
20	0.342	.....	.....	0.10	0.11	0.417	0.029	.....	.....	.....	.....	.....	.....	.....	W. B. Kunhardt	B. & M.	Ingot 6 × 6 and rolled bar ¾" diam.
21	0.46	.....	.....	0.369	1.06	0.055	0.560	.....	.....	.....	.....	.....	.....	.....	A. Wahlberg	.....	Rolled bar 0.7" square.
II. Hyper-eutectoid steel																	
22	1.143	.....	.....	0.124	0.24	0.016	0.009	147,580 (103.8)	89,455 (62.9)	84,920 (59.7)	5.0	3.6	0.050	277	J. A. Mathews	B. & M.	Rolled bar ⅝ in. diam.
23	1.2	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	W. Campbell	.....	Rolled bar 1 in. diam.
24	1.45	.....	.....	0.14	0.16	0.009	0.006	150,200 (105.6	85,460 (60.1)	81,560 (57.3)	5.7	8.9	0.014	293	J. A. Mathews	B. & M.	Rolled bar ⅝ in. diam.
25	1.72	.....	.....	0.20	0.075	0.013	0.018	.....	.....	.....	.....	.....	.....	.....	W. Campbell	M.	Rolled bar ¼" square.
25.1	2.04	.....	.....	.....	0.28	0.014	0.014	.....	.....	.....	.....	.....	.....	.....	" "	M.	Rolled bar ¼" square.
III. Cast irons																	
26	2.954	1.618	1.356	0.040	Nil	0.050	0.035	.....	.....	.....	.....	.....	.....	.....	Columbia[b]	B.	Buttons about 2 in. diam.
27	4±	1.3±	2.7±	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....			
28	3.055	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....		B.	Thin plates poured on iron plates.
29	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....			
30	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....			
31	3.5	3.5	Nil	.....	.....	0.01	0.015	.....	.....	.....	.....	.....	.....	.....	W. Campbell		Ingot about 5 × 5 in. × more than 8 in.
32	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	R. Moldenke		Cast bar ½ in. square.
33	3.16	0.50	2.66	1.82	0.49	0.504	0.061	.....	.....	.....	.....	.....	.....	.....	W. A. Bole	M.	Inverted pyramid 4 × 4 and 1½ × 1½, about 12 in. long.
34	3.978	.....	.....	0.89	0.84	0.180	0.026	.....	.....	.....	.....	.....	.....	.....	R. H. Sweetser	B. & M.	Kish = yielding iron ingot about 10½ lb. weight.
35	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	Anonymous	.....	Piece of radiator, thin.
36	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	"	.....	" " " thick.
37	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	A. W. Whitney	.....	Tread of chilled and plate cast-iron
															" " "		Plate car wheel
38	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....		.....	Spiegeleisen.
39	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....		.....	Müsen Spiegeleisen.
40	5±	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	H. B. Weaver	.....	Half ingot, 52 lb. weight.
42	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....			
IV. Wrought iron and miscellaneous																	
50	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	Anonymous	.....	Wrought-iron bar 1 in. × ½ in.
																.....	Muck bar.
51	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....		.....	Medium C steel casting in form of runner, 2 in. diam.
52	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	Taylor Wharton Iron and Steel Co.	M.	Cast bars 12 in. × 1 in. × ½ in.
60	1.24	.....	.....	.....	12.8	.....	.....	.....	.....	.....	.....	.....	.....	.....	Goldschmidt Thermit Co.	B.	Cast bars 6 in. × 1 in. × ½ in.
61	.....	.....	Nil	.....	33.0	.....	.....	.....	.....	.....	.....	.....	.....	.....	R. H. Illingworth	M.	Rolled bar ½" diam.
65	0.20±[c]	.....	36±[c]	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	Anonymous	C. Mills	Rolled bar 1" square.
66	0.29	.....	4.61	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....			
															W. E. Ruder	M.	Flat bars 2 in. × ¼ in.
70	0.050[c]	.....	.....	4[c]	0.072[c]	0.010[c]	0.028[c]	.....	.....	.....	.....	.....	.....	.....			Sheets 0.022" thick.
															" " "	M.	Sheets 0.015 " thick.
71	.....	.....	.....	2.5	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....			

[a] B = Booth, Garrett and Blair; M = makers of the metal; P. T. L. = Pittsburg Testing Laboratory.

[b] These represent small lots of cast iron made by melting Burgess pure electrolytic iron, No. 1, in the metallurgical laboratories of Columbia University. Nos. 26 and 27 were made from this electrolytic iron alone, melted in graphite crucibles brasqued with cane sugar charcoal. Nos. 28, 29, and 30 were made from this same iron diluted with additions of steel No. 24.

[c] Typical composition.

N.B.—Not broken.

TABLE 39.—COMPOSITION, PROPERTIES, AND SOURCES OF THE STEELS AND IRONS USED IN THE EXPERIMENTS MADE FOR THIS WORK

Designation number	Chemical composition, per cent.							Mechanical properties, after coarsening at 1,100°, grain refining, and annealing 3 hours at 600° (except No. 0, as rolled, and No. 2, annealed 3 hours at 550°)							Source	Analyzed by[a]	Form of specimens
	Carbon			Si	Mn	P	S	Tensile strength, lb. per sq. in. (kg. per sq. mm.)	Yield point, lb. per sq. in. (kg. per sq.mm.)	Proportionality limit, lb. per sq. in. (kg. per sq. mm.)	Elongation in 4 in., per cent.	Contraction of area, per cent.	Fremont test, kg. m. per sq. mm.	Brinell hardness No.			
	Total	Combined	Graphitic														
I. Electrolytic iron, hypo-eutectoid and eutectoid steels, A, normal																	
0	0.01	.....	.....	Very low				44,190 (31.1)	20,180 (14.2)	.....	27.8	68.6	.....	75	Am. Rolling Mill Co.	.....	Billet about 4 in. square.
1	0.023	.....	.....	.....	Nil	.....	.....	.....	.....	.....	.....	.....	.....	.....	C. F. Burgess	B.	Cathode sheet.
2	0.027	.....	.....	0.003	0.26	0.005	0.024	51,040 (35.9)	39,535 (27.8)	38,850 (27.3)	31.5	77.0	0.50 N.B.	92	W. B. Kunhardt	B. & M.	Rolled bar ⅝ in. diam.
2.1	0.05	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	C. S. Marshall & F. H. Daniels	M.	Rolled bar ⅝ in. diam.
3	0.105	.....	.....	0.013	0.24	0.015	0.028	.....	.....	.....	.....	.....	.....	.....	Pittsburg T. L.	P. T. L.	Plate 1 by 5/16 in.
4	0.214	.....	.....	0.039	0.05	0.013	0.010	80,600 (56.7)	61,390 (43.2)	59,900 (42.1)	12.0	68.5	0.36 N.B.	153	W. B. Kunhardt	B. & M.	Rolled bar ⅝ in. diam.
5	0.227	.....	.....	0.039	0.05	0.013	0.010	.....	.....	.....	.....	.....	.....	.....	" " "	B. & M.	Rolled bar ⅝″ diam.
6	0.235	.....	.....	0.039	0.05	0.013	0.010	.....	.....	.....	.....	.....	.....	.....	" " "	B. & M.	Rolled bar ⅝″ diam.
7	0.244	.....	.....	0.050	Nil	Nil	0.007	.....	.....	.....	.....	.....	.....	.....	C. F. Burgess	B.	Rolled bar ⅝″ diam.
8	0.382	.....	.....	0.027	0.22	.....	0.004	.....	.....	.....	.....	.....	.....	.....	W. Campbell	B. & R. G. Morse	Rolled bar ¾ in. diam.
9	0.40	.....	.....	0.103	0.16	0.014	0.012	88,290 (62.1)	65,275 (45.9)	57,800 (40.6)	15.5	57.7	0.203	187	W. B. Kunhardt	B.	Rolled bar ⅝ in. diam.
10	0.563	.....	.....	0.18	0.15	0.013	0.013	.....	.....	.....	.....	.....	.....	.....	J. A. Mathews	B. & M.	Rolled bar ½″ diam.
11	0.59	.....	.....	0.144	0.17	0.018	0.013	109,840 (77.2)	74,000 (52.0)	69,650 (49.0)	15.8	49.8	0.056	202	C. S. Marshall & F. H. Daniels	B. & M.	Rolled bar ⅝ in. diam.
12	0.73	.....	.....	0.141	0.07	0.012	0.019	.....	.....	.....	.....	.....	.....	.....	W. Campbell	B. & M.	Rolled bar 5/16″ square.
13	0.78	.....	.....	0.108	0.26	0.011	0.013	.....	.....	.....	.....	.....	.....	.....	Pittsburg T. L.	P. T. L.	Plate ⅛ in. thick.
14	0.92	.....	.....	0.14	0.123	0.009	0.011	109,840 (77.2)	74,340 (52.3)	.....	3.8	13.7	0.086	212 223	J. A. Mathews	B. & M.	Plate ⅝ in. diam.
B. High manganese																	
15	0.11	.....	.....	.....	0.67	.....	.....	.....	.....	.....	.....	.....	.....	.....	Dealer	P. T. L.	Rolled bar 1″ square.
16	0.110	.....	.....	0.004	0.60	0.036	0.064	.....	.....	.....	.....	.....	.....	.....	W. Campbell	B.	Rolled bar ½″ square.
17	0.140	.....	.....	0.040	1.16	0.020	0.018	78,130 (54.9)	60,160 (42.3)	.....	13.8	73.6	0.504 N.B.	126	W. B. Kunhardt	B. & M.	Rolled bar ⅝ in. square.
17.1	0.147	.....	.....	0.018	0.655	.....	.....	.....	.....	.....	.....	.....	.....	.....	Dealer	B.	Cold-drawn ¾″ diam.
18	0.32	.....	.....	0.122	0.41	0.004	0.022	.....	.....	.....	.....	.....	.....	.....	Pittsburg T. L.	P. T. L.	Forging ½ in. diam.
19	0.46	.....	.....	0.094	1.215	0.017	0.015	122,720 (86.3)	107,850 (75.8)	101,210 (71.2)	11.8	56.0	0.201	192	W. B. Kunhardt	B.	Rolled bar ⅝ in. diam.
19.1	0.46	.....	.....	.....	0.466	0.041	.....	.....	.....	.....	.....	.....	.....	.....	Anonymous	B.	Rolled bar 1″ square.
C. High phosphorus and high sulphur																	
20	0.342	.....	.....	0.10	0.11	0.417	0.029	.....	.....	.....	.....	.....	.....	.....	W. B. Kunhardt	B. & M.	Ingot 6 × 6 and rolled bar ¾″ diam.
21	0.46	.....	.....	0.369	1.06	0.055	0.560	.....	.....	.....	.....	.....	.....	.....	A. Wahlberg	.....	Rolled bar 0.7″ square.
II. Hyper-eutectoid steel																	
22	1.143	.....	.....	0.124	0.24	0.016	0.009	147,580 (103.8)	89,455 (62.9)	84,920 (59.7)	5.0	3.6	0.050	277	J. A. Mathews	B. & M.	Rolled bar ⅝ in. diam.
23	1.2	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	W. Campbell	.....	Rolled bar 1 in. diam.
24	1.45	.....	.....	0.14	0.16	0.009	0.006	150,200 (105.6	85,460 (60.1)	81,560 (57.3)	5.7	8.9	0.014	293	J. A. Mathews	B. & M.	Rolled bar ⅝ in. diam.
25	1.72	.....	.....	0.20	0.075	0.013	0.018	.....	.....	.....	.....	.....	.....	.....	W. Campbell	M.	Rolled bar ¼″ square.
25.1	2.04	.....	.....	.....	0.28	0.014	0.014	.....	.....	.....	.....	.....	.....	.....	" "	M.	Rolled bar ¼″ square.
III. Cast irons																	
26	2.954	1.618	1.356	0.040	Nil	0.050	0.035	.....	.....	.....	.....	.....	.....	.....	Columbia[b]	B.	Buttons about 2 in. diam.
27	4±	1.3±	2.7±	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....			
28	3.055	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....		B.	Thin plates poured on iron plates.
29	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....			
30	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....			
31	3.5	3.5	Nil	.....	.....	0.01	0.015	.....	.....	.....	.....	.....	.....	.....	W. Campbell		Ingot about 5 × 5 in. × more than 8 in.
32	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	R. Moldenke		Cast bar ½ in. square.
33	3.16	0.50	2.66	1.82	0.49	0.504	0.061	.....	.....	.....	.....	.....	.....	.....	W. A. Bole	M.	Inverted pyramid 4 × 4 and 1½ × 1½, about 12 in. long.
34	3.978	.....	.....	0.89	0.84	0.180	0.026	.....	.....	.....	.....	.....	.....	.....	R. H. Sweetser	B. & M.	Kish = yielding iron ingot about 10½ lb. weight.
35	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	Anonymous	.....	Piece of radiator, thin.
36	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	" " "	.....	" " " thick.
37	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	A. W. Whitney	.....	Tread and plate of chilled cast-iron car wheel
38	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	" " "		
39	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	Spiegeleisen.
40	5±	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	Müsen Spiegeleisen.
42	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	H. B. Weaver	.....	Half ingot, 52 lb. weight.
IV. Wrought iron and miscellaneous																	
50	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	Anonymous	.....	Wrought-iron bar 1 in. × ½ in.
51	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	Muck bar.
52	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	Taylor Wharton Iron and Steel Co.	.....	Medium C steel casting in form of runner, 2 in. diam.
60	1.24	.....	.....	.....	12.8	.....	.....	.....	.....	.....	.....	.....	.....	.....		M.	Cast bars 12 in. × 1 in. × ½ in.
61	.....	.....	Nil	.....	33.0	.....	.....	.....	.....	.....	.....	.....	.....	.....	Goldschmidt Thermit Co.	B.	Cast bars 6 in. × 1 in. × ½ in.
65	0.20±[c]	.....	36±[c]	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	R. H. Illingworth	M.	Rolled bar ½″ diam.
66	0.29	.....	4.61	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	Anonymous	C. Mills	Rolled bar 1″ square.
70	0.050[c]	.....	.....	4[c]	0.072[c]	0.010[c]	0.028[c]	.....	.....	.....	.....	.....	.....	.....	W. E. Ruder	M.	Flat bars 2 in. × ¼ in.
71	.....	.....	.....	2.5	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	" " "	M.	Sheets 0.022″ thick.
																	Sheets 0.015″ thick.

[a] B = Booth, Garrett and Blair; M = makers of the metal; P. T. L. = Pittsburg Testing Laboratory.

[b] These represent small lots of cast iron made by melting Burgess pure electrolytic iron, No. 1, in the metallurgical laboratories of Columbia University. Nos. 26 and 27 were made from this electrolytic iron alone, melted in graphite crucibles brasqued with cane sugar charcoal. Nos. 28, 29, and 30 were made from this same iron diluted with additions of steel No. 24.

[c] Typical composition.

N.B.—Not broken.

844. Other Miscellaneous Tables

TABLE 40.—COMPOSITION, DENSITY, AND HARDNESS OF CERTAIN METARALS, EUTECTICS, AND EUTECTOIDS

	Substance	Composition			Specific gravity	Hardness, Brinell No.	Hardness, Moh's scale
		Fe, per cent.	C, per cent.	P, per cent.			
1	Pearlite, the iron-cementite entectoid, 86.5 per cent. Fe, 13.5 Fe_3C.	99.10	0.90	...	7.846[1] 7.827[2]	166[1] 207[1] 199[3]	....
2	Ledeburite, the austenite-cementite eutectic of white cast-iron, 41.0 per cent. pearlite, 59.0 per cent. Fe_3C.	95.70	4.30	.....	.....	.....	.....
3	Cementite, Fe_3C, in steels with less than 1.26 per cent. C.	93.33	6.67	.....	7.74[3]	.....	.....
	Cementite, Fe_3C, in steels with more than 1.26 per cent. C.	93.33	6.67	.....	7.24[3]	.....	.....
4	Richest non-eutectiferous steel, 74.5 per cent. Fe, 25.5 per cent. Fe_3C.	98.30	1.70	.....	7.798[3]	185[3]	.....
5	Graphite	.....	100.0	.....	2.255[4]	.....	.....
6	Iron (A.R.M. steel) at 15°	99.95±	0.01	.....	7.870[1]	75[1]	3.5[7]
7	Iron pure	.....	.....	.....	7.85–7.88[5]	.....	.....
8	Iron pure, by extrapolation 15°/4°	.....	.....	.....	7.85±[3]	80±[3]	.....
9	Iron phosphide Fe_3P	84.42	.....	15.58	6.74[6]	.....	6.5[7]
10	Iron phosphide Fe_2P	78.32	.....	21.68	6.56[6]	.....	.....
11	Iron phosphide FeP	64.37	.....	35.63	5.76[6]	.....	.....
12	Iron phosphide, Fe_2P_3	54.63	.....	45.37	4.5[6]	.....	.....
13	Ternary Fe-C-P eutectic	91.2	1.9	6.9	.....	.....	.....
14	Steadite (Sauveur), iron-phosphorus eutectic, a conglomerate of (1) iron containing 1.7 per cent. P and (2) Fe_3P...	90.0	.....	10.0	.....	.....	.....

[1] Author, all the Author's specific gravities are with both substance and water at 15°C.

[2] Benedicks, *Journ. Iron and Steel Inst.*, 1908, No. II, vol. 77, p. 221.

[3] Benedicks, "Recherches Phys. et Phys. Chim. sur l'Acier au Carbone," Doctorate Thesis. Upsala, 1904, pp. 31, 35, and 95. Benedicks' samples were at room temperature (14 to 20°C.) compared with water at 4°C. In the determination of the density of iron, line 8, the observations were made at 15°, and the observed specific gravity was then reduced to terms of that of water at 4°.

[4] Le Chatelier and Wologdine, *Compt. rendus*, 1908, vol. 146, p. 53.

[5] Landolt, Börnstein, and Roth, "Physikalisch-Chemische Tabellen," Berlin, 1912.

[6] Le Chatelier and Wologdine, *Compt. rendus*, 1909, vol. 149, p. 709.

[7] Saklatwalla, *Journ. Iron and Steel Inst.*, 1908, No. II, vol. 77, p. 103. "The smallest amount of phosphorus" raises the hardness of iron to between 5 and 5.5. This refers to a phosphorus content estimated at between 0.10 and 0.30 per cent. (*Idem*, private communication, Dec. 24, 1915).

TABLE 40A.—CERTAIN STOICHIOMETRIC DATA

1. 1 per cent. Si increases specific volume of iron by about 0.00117 c.c.[1]
2. 1 per cent. carbon represents 15.0 per cent. cementite Fe_3C, containing 6.67 per cent. C.
   1 " " silicon " 3.0 " " iron silicide FeS.
   1 " " phosphorus " 6.4 " " iron phosphide Fe_3P
   1 " " phosphorus " 10.0 " " steadite ($Fe + Fe_3P$) eutectic.
   1 " " sulphur " 2.75 " " ferrous sulphide, FeS.
   1 " " sulphur " 2.72 " " manganese sulphide MnS.
3. The eutectic ledeburite, if it contains 4.30 per cent. of carbon, and if it cools slowly without graphitizing, will then contain:

[1] Benedicks, Doctorate Thesis, Upsala, 1904, p. 32.

	By weight	By volume, ±
Pearlite	41.04	39.1
Eutectic cementite	52.35	54.1
Pro-eutectoid cementite	6.61	6.8
Total	100.00	100.00
Austenoid = pearlite + pro-eutectoid cementite	47.65	60.9

4. Incomplete graphitization.

Fe_3C = Gr + Austenite of 1.7 per cent. C..		By weight	By volume, ±	Containing by weight Carbon	Iron
100 parts Fe_3C yield:	Graphite	5.05	15.47	5.05	0
	Pearlite	81.66	71.90	0.73	80.93
	Pro-eutectoid Fe_3C.	13.29	12.63	0.89	12.40
	Total	100.00	100.00	6.67	93.33

In the products of this graphitization there are 16.2 parts of pearlite by weight, and 4.6± parts by volume, for each part of graphite. This, of course takes no account of the products of the transformation of any primary or eutectic austenite accompanying the initial cementite.

5. Complete graphitization.

Fe_3C = Gr + 3Fe, *i.e.*, the pearlitic and pro-eutectoid Fe_3C both graphitize.

		By weight	Containing carbon	By volume, ±
100 parts Fe_3C yield:	Graphite	6.67	6.67	20
	Ferrite	93.33	0.00	80
	Total	100.00	6.67	100

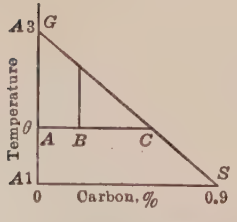
In the products of this graphitization there are 14 parts of ferrite by weight and 4± parts by volume for each part of graphite. This of course takes no account of the products of the transformation of any primary or eutectic austenite accompanying the initial cementite.

6. In cast iron, formulæ[1] for per cent. ferrite (Fe), cementite (Cem.), pearlite (Pe), graphite (Gr), and combined carbon (C. C.),

$$\text{if C.C.} < 0.90 \begin{cases} \text{Fe} + \text{Pe} + \text{Gr} = 100 \\ \text{Pe} = 111\ \text{C.C.} \end{cases}$$

$$\text{if C.C.} > 0.90 \begin{cases} \text{Pe} + \text{Cem.} + \text{Gr} = 100 \\ \text{Pe} = 1.156(100 - \text{G} - 15\text{C}) \end{cases}$$

7. Percentage of pro-eutectoid ferrite due in hypo-eutectoid steel at temperature θ within the transformation range. Let abscissa of curve GS at θ, = present carbon content of the austenite = AC; carbon content of the steel = AB per cent.



$$\text{Per cent. of austenite by weight} = \frac{100 \times AB}{AC}$$

$$\text{Per cent. of pro-eutectoid ferrite} = 100 - \frac{100 \times AB}{AC} = \frac{100 \times BC}{AC}$$

If the steel is quenched from θ, and if the whole of the austenite is caught as martensite, these formulæ will give the percentage of pro-eutectoid ferrite on one hand, and on the other hand that of martensite, and of its decomposition products, troostite and sorbite, formed on reheating.

[1] Modified from Sauveur, "Metallography of Iron and Steel," Sauveur and Boylston, Cambridge, Mass., 1912, Lesson 19, p. 11.

TABLE 41.—CERTAIN CRITICAL TEMPERATURES

		0.0 per cent. carbon	0.48 per cent. carbon	0.90 per cent. carbon
		Temperature, °C.		
A1	Ac1	.....	738[5]	761,[4] 730,[7] 737[8]
	Ae1	.....	725[6]	725
	Ar1	.....	703, 715[5]	707,[4] 711,[7] 711,[8] 699[9]
A2	Ae2	768[1]	768	768
A3	Ac3	935[2]	813[4]	761[4]
	Ae3	917[3]	769[3]	725[3]
	Ar3	904[2]	748[4]	707[4]

Ae3 for steel with 0 to 0.483 per cent. C =				$917^{0} - 306 \times$ carbon per cent.[3]
Ae3 for steel with 0.483 to 0.90 per cent. C =				$820^{0} - 105.5$ carbon per cent.[3]
Austenite-cementite eutectic, ledeburite				Melts 1135[10]
Osmondite stage at which troostite becomes sorbite fully				400[11]
Sorbite becomes granular pearlite in Heyn's eutectoid steel				640[12]
Iron—iron phosphide eutectic, steadite, 10.2 per cent. P, or 65.3 per cent. Fe_3P + 34.7 per cent. Fe				Melts 1,007,[13] 980,[14] 980[15]
Iron—iron phosphide eutectic, easily undercooled to				880,[14] 944[15]
Ternary Fe-C-P eutectic,	Fe 91.3 per cent.,	C 2.0 per cent.,	P 6.7 per cent.	Melts 950[16]
	91.15	1.96	6.86	Melts 953[17]
	91.19	1.92	6.89	Melts 945[18]
Nickel, critical range,				320,[19] 250 to 375[20]
Magnesite, $MgCO_3$, dissociation point				650[21]
Cupric sulphate, $CuSO_4$, decomposes spontaneously				655[22]
Calcite, $CaCO_3$, dissociation point				850[21]
Silver sulphate, Ag_2SO_4 decomposes spontaneously				870[22]
Cobalt, critical range				1,159,[19] 1,000 to 1,080[20]
Ferrous sulphide, FeS, 63.6 per cent. Fe, 36.4 per cent. S				Melts 1,188,[25] 1,171 ± 10[26]
Manganese sulphide, MnS, 63.4 per cent. Mn, 36.6 per cent. S				Melts 1,620[25]
Eutectic of 7 per cent. MnS + 93 per cent. FeS, or 59.2 per cent. Fe, 4.4 per cent. Mn, 26.4 per cent. S				Melts 1,182[25]
Up to 13 per cent. Si, 1 per cent. of Si lowers the combined carbon in cast iron from 4.3 by 0.284 per cent., or				Comb. carbon = $4.3 - 0.284 \times$ silicon per cent.[23]
Correction for Osmond's early data				$\theta = \theta' \times 1.048 - 6$[24]
θ = corrected temperature, θ' = uncorrected temperature				Thus 750° becomes 780°; 855° becomes 890°

References to Table 41.—[1] Burgess, G. K., and Crowe, *Trans. Amer. Inst. Min. Eng.*, 1913, vol. 47, p. 699. [2] Burgess, G. K., quoted by the Author, *idem*, p. 645. [3] Author, *idem*, p. 647. Howe and Levy, *idem*, p. 592. [4] Burgess, G. K., Crowe, and Rawdon, *idem*, p. 606. [5] Howe and Levy, unpublished results for steel with 0.463 per cent. C. [6] Author, *Trans. Amer. Inst. Min. Eng.*, 1913, vol. 47, pp. 647 and 743. [7] Stead, *Journ. Iron and Steel Inst.*, 1913, No. II, vol. 88, p. 409. Stead's "thermal profile method" gave 728° to 731° for Ac1, and 710° to 712° for Ar1. [8] *Idem*, pp. 407 and 411. The thermal values found by 16 observers on Stead's steel gave values ranging from 719° to 746° (nine cases between 731° and 742°, average 737°) for Ac1, and from 697° to 720° (eight cases between 705° and 716°, average 711°) for Ar1. [9] Carpenter and Keeling, *idem*, 1904, No. I, vol. 65, p. 233. [10] Gutowsky, *Metallurgie*, 1909, vol. 6, p. 731. [11] Heyn and Bauer, *Stahl u. Eisen*, 1906, vol. 26, p. 996. [12] *Idem*, p. 784. [13] Saklatwalla, *Journ. Iron and Steel Inst.*, 1908, No. II, vol. 77, p. 92; and *Metallurgie*, 1908, vol. 5, p. 711. [14] Gercke, *idem*, p. 604. [15] Gutowsky, *idem*, p. 470. [16] Wüst, *idem*, p. 87. [17] Goerens and Dobbelstein, *idem*, p. 561. [18] Stead, "Presidential Address," *Chem. Section British Assoc.*, 1910; and *Journ. Iron and Steel Inst.*, 1915, No. I, vol. 91, p. 141. [19] Guertler, *Metallographie*, Borntraeger, Berlin, 1912, vol. 1, Part 1, pp. 71 and 73. [20] Honda, *Science Reports*, Tohoku University, vol. 2, No. 2, Sendai, Japan, 1913, pp. 75 and 76. [21] Le Chatelier, "Sur la mesure des températures elevées developpées dans les foyers industries." Prior to 1902. [22] Bradford, R. H., *Trans. Amer. Inst. Min. Eng.*, 1903, vol. 23, p. 50. [23] Wüst and Peterson, *Metallurgie*, 1906, vol. 3, p. 815. [24] Osmond, *Journ. Iron and Steel Inst.*, 1906, No. IV, vol. 71, p. 451. [25] Röhl, *Iron and Steel Inst., Carnegie Mem.*, 1912, vol. 4, p. 52. [26] Friedrich, *Metallurgie*, 1908, vol. 5, p. 56.

TABLE 42.—INDUSTRIAL AND MISCELLANEOUS TEMPERATURES

	Temperature	
	°C.	°F.
Liquid air boils at (see Table 44)	−192 to −182	−314 to −296
CO begins to reduce precipitated Fe_2O_3 at about[1]	141	286
CO begins to reduce Cleveland ore[1]	199	390
CO begins to deposit carbon between[1]	200 and 221	392 and 430
Deposited carbon begins to react on iron oxide at or below[1]	249	480
CO_2 begins to oxidize metallic iron between[1]	299 and 417	570 and 783
Carbon deposition is most rapid between[1]	400 and 450	752 and 842
CO_2 oxidizes soft but not hard coke[1]	417	783
CO_2 begins to oxidize spongy iron at or below[1]	417	783
Chimney flue temperature, basic open hearth[2]	500 to 600	932 and 1,112
Air-gas regenerator temperature, basic open hearth[2]	1,000 to 1,200	1,832 and 2,192
Rail finishing temperature, usually[3]	935 ± 50	1,715 ± 90
Rail-ingot rolling temperature[3]	1,075 to 1,150	1,967 and 2,102
Finishing temperature for boiler plates about ⅜ in. thick	About 700	About 1,292
Scales just form but do not fall away when cooling in air[4]	843	1,550
Free scaling temperature[4]	899	1,650

[1] Bell, cited by the Author, "Metallurgy of Steel," N. Y., 1891, p. 118.

[2] Simmersbach, *Stahl u. Eisen*, 1911, vol. 31, pp. 1996, 1680.

[3] Burgess, Crowe, Rawdon and Waltenberg, *Technological Paper No.* 38, Bur. of Standards, Washington, D. C., 1914; *Iron Age*, 1914, vol. 94, p. 837; *Bull. Amer. Inst. Min. Eng.*, Sept., 1914, p. 2433.

[4] White and Taylor, *Trans. Amer. Soc. Mech. Engin.*, 1900, vol. 21, p. 628.

TABLE 43.—TEMPER COLORS AND COLOR NAMES FOR VARIOUS TEMPERATURES

	Degrees C.	Degrees F.
Light straw	149[1]	300
Pale yellow	220[2]–225[3]	428–437
Straw	204[1]–230[2]	400–446
Golden yellow	243[2]–250[3]	469–482
Yellow	250[4]	482
Brown	255[2]–275[3]	491–527
Brown dappled with purple	265[2]	509
Purple	277[2]	531
Purple bronze approaching blue	260[1]–316[1]	500–600
Violet	300[5]–325[3]	572–617
Blue	288[2], 300[4]–350[3,5]	550, 572–662
Dark blue	316[2]–371[1]	600–700
Blue gray	400[3,5]	752
Gray	450[4]	842
Lowest red visible in the dark	480[6]	896
"Black heat," dark blood red, black red	532[7]	990
Dark cherry red	635[7]	1175
Cherry full red	746[7]	1375
Light cherry, bright cherry, light red	843[7]	1550
Orange	899[7]	1650
Yellow	996[7]	1825
Light yellow	1080[7]	1975
White	1152[7]–1204	2105–2200

Notes to Table 43.—[1] Howard, *Report Chief Ordnance U. S. Army*, 1893, p. 703.

[2] Ede, "Treatment of Steel," Miller, Metcalf, and Parkin; the Author, "Metallurgy of Steel," 1891, p. 23.

[3] Guillet and Portevin, *Rev. Metallurgie, Memoires*, 1909, vol. 6, p. 102.

[4] Maurer, *idem*, 1908, vol. 5, p. 716.

[5] Goerens and Hartel, *Zeits. anorg. Chem.*, 1913, vol. 81, p. 143.

[6] Robin, *Bull. Soc. d'Encouragement*, 1912, Part 2, p. 219.

[7] White and Taylor, *Trans. Amer. Soc. Mech. Engineers*, 1900, vol. 21, p. 628.

TABLE 44.—MELTING AND BOILING POINTS

	Melting point, °C.	Boiling point, °C.	Melting point, °F.	Boiling point, °F.
Absolute zero[1]	−273.1		−459.6	
Hydrogen[2,3]	−259	−252.5	−434	−422.5
Fluorine[2,4]	−223	−187	−369	−305
Oxygen[2,3]	−218	−183	−360	−297.4
Nitrogen[2,3]	−210	−195.6	−346	−320.1
Liquid air[4]	.....	−192 to −182	.....	−314 to −296
Mercury[2,3]	− 38.7	357	− 37.7	675
Water	0	100	32	212
Phosphorus[2,3]	44	111.2	111.4	232.5
Potassium[2]	62.3	144.1	144.0	291
Sodium[2]	97.5	.....	207.5	.....
Sulphur[2,3] S_1	235.0	.....	112.8	.....
Sulphur[2,3] S_2	246.6	444.5	119.2	832.1
Sulphur[2,3] S_3	224.2	.....	106.8	.....
Naphthalene[3]	.....	218	.....	424.4
Selenium[2]	217–220	.....	422–428	.....
Tin[2]	231.9	.....	449.4	.....
Bismuth[2]	271	.....	520	.....
Cadmium[2]	320.9	.....	609.6	.....
Lead[2]	327.4	.....	621.1	.....
Zinc[2,3]	419.4	918	786.9	1684
Antimony[2]	630.0	.....	1166	.....
Aluminum[2]	658.7	.....	1217.7	.....
Bronze, 20 per cent. tin, 80 per cent. copper[14]	790	.....	1454	.....
Sodium chloride[5]	800	.....	1472	.....
Silver, in air[6]	955	.....	1751	.....
Silver in reducing atmos.[2]	960.5	.....	1761	.....
Gold[2]	1063	.....	1945.5	.....
Copper in air[6]	1062	.....	1943.6	.....
Copper in reducing atmos.[2]	1083	.....	1981.5	.....

	Melting point, °C.	Melting point, °F.
Potassium sulphate[8]	1060	1940
Cuprous sulphide[7]	1135±10	2075±18
Eutectic of 7 per cent. MnS + 93 per cent. FeS[10]	1182	2160
Ferrous sulphide[10]	1188	2170
Ferrous sulphide[7]	1171±10	2140±18
Manganese	1225	2237
$Fe_3Mn_2S_8$[10]	1362	2484
Ferrous oxide, FeO[9]	1419	2586
Nickel[2]	1452	2646
Cobalt[2]	1490	2714
Chromium[2]	1510	2750
Iron[8]	1530	2786
Magnetic oxide,[9] Fe_3O_4	1538	2800
Magnetic oxide,[12] Fe_3O_4	1527	2781
Palladium[2]	1549	2820
Ferric oxide,[12] Fe_2O_3*	1565	2849
Manganese sulphide, MnS[10]	1620	2948
Platinum[2]	1755	3191
Eutectic of Mg_2SiO_4 + MgO[11]	1850±20	3362±36
Mg_2SiO_4[11]	1890±20	3434±36
Calcium oxide[9]	1995	3623
Alumina[9]	2020	3668
Iridium[2]	2300?	4172?
Vanadium carbide[9]	2750	4982
Tungsten[2]	3000	5432
*Ferric oxide[13] Fe_2O_3, sinters at	1360	2480

References to Table 44.—[1] H. L. Callendar, "Encyclopædia Britannica," 1911, vol. 26, p. 829.

[2] "Melting Points of the Chemical Elements," Circular No. 35 of the Bureau of Standards, Washington, D. C., June 15, 1912, p. 2.

[3] Landolt, Börnstein, and Roth; "Physikalisch-Chemische Tabellen," Berlin, 1912, pp. 206, 324, 373 *et seq.*

[4] J. Dewar, "Encyclopædia Britannica," 1911, vol. 16, p. 750.

[5] Burgess-Le Chatelier, "Measurement of High Temperatures," Wiley, New York, 1912; pp. 7, 189, 190, and 434.

[6] J. A. Harker, "Technical Thermometry," Cambridge (England) Scientific Instrument Co., Ltd., 1906, p. 56.

[7] Friedrich, *Metallurgie*, 1908, vol. 5, p. 56.

[8] G. K. Burgess, private communication, April 7, 1914.

[9] Ruff, *Stahl u. Eisen*, vol. 31, p. 1194.

[10] Röhl, *Iron and Steel Inst., Carnegie Memoirs*, 1912, vol. 4, p. 52.

[11] Bowen and Anderson, *Amer. Journ. Sci.*, 1914, vol. 37, p. 487.

[12] Kohlmeyer, *Metallurgie*, 1909, vol. 6, p. 325.

[13] Hofman and Mostowitsch, *Trans. Amer. Inst. Min. Eng.*, 1909, vol. 40, p. 808. In dry air Fe_2O_3 sinters at 1,360°C. and loses 4.4 per cent. oxygen, becoming magnetic at 1,375°.

[14] Le Chatelier, "Sur la mesure des températures élevées developpées dans les foyers industries." Prior to 1902.

TABLE 45.—COMPARISON OF THERMOMETRIC SCALES, FOR EVERY 10°C. AND EVERY 25°F.

C.	F.	C.	F.	C.	F.	C.	F.	C.	F.	C.	F.
−273.1	−459.6	300	572	630	1166	866	1591	1163	2125	1496	2725
Absolute zero		302	575	635	1175	870	1598	1170	2138	1500	2732
−200	−328	310	590	640	1184	871	1600	1177	2150	1510	2750
−100	−148	316	600	649	1200	880	1616	1180	2156	1520	2768
− 40	− 40	320	608	650	1202	882	1620	1190	2174	1524	2775
−17.7	0	330	625	657	1215	885	1625	1191	2175	1530	2786
− 4	+ 25	330	626	660	1220	890	1634	1200	2192	1538	2800
0	32	340	644	663	1225	897	1647	1205	2200	1540	2804
+ 10	50	343	650	664	1227	899	1650	1210	2210	1550	2822
20	68	350	662	670	1238	900	1652	1219	2225	1552	2825
24	75	357	675	677	1250	910	1670	1220	2228	1560	2840
30	86	360	680	678	1252	913	1675	1230	2246	1566	2850
38	100	370	698	680	1256	920	1688	1232	2250	1570	2858
40	104	371	700	690	1274	927	1700	1240	2264	1580	2875
50	122	380	716	691	1275	930	1706	1246	2275	1580	2876
52	125	385	725	693	1280	940	1724	1250	2282	1590	2894
60	140	390	734	700	1292	941	1725	1260	2300	1593	2900
66	150	399	750	705	1300	950	1742	1270	2318	1600	2912
70	158	400	752	707	1304	952	1746	1274	2325	1625	2957
79	175	410	770	709	1308	955	1750	1280	2336	1650	3002
80	176	413	775	710	1310	960	1760	1288	2350	1675	3047
90	194	420	788	714	1317	969	1775	1290	2354	1700	3092
93	200	427	800	719	1325	970	1778	1300	2372	1750	3182
100	212	430	806	720	1328	973	1783	1302	2375	1800	3272
107	225	440	824	722	1332	980	1796	1310	2390	1850	3362
110	230	441	825	726	1338	982	1800	1316	2400	1900	3452
120	248	450	842	730	1346	984	1803	1320	2408	1950	3542
121	250	455	850	732	1350	990	1814	1330	2425	2000	3632
130	266	460	860	736	1357	996	1825	1330	2426	2050	3722
135	275	469	875	737	1359	1000	1832	1340	2444	2100	3812
140	284	470	878	740	1364	1010	1850	1343	2450		
149	300	480	896	746	1375	1020	1868	1350	2462		
150	302	482	900	750	1382	1024	1875	1357	2475		
160	320	490	914	756	1393	1030	1886	1360	2480		
163	325	496	925	760	1400	1038	1900	1370	2498		
170	338	500	932	770	1418	1040	1904	1371	2500		
177	350	510	950	774	1425	1050	1922	1380	2516		
180	356	520	968	779	1434	1052	1925	1385	2525		
190	374	524	975	780	1436	1060	1940	1390	2534		
191	375	530	986	788	1450	1066	1950	1399	2550		
200	392	538	1000	790	1454	1070	1958	1400	2552		
204	400	540	1004	794	1461	1080	1975	1410	2570		
210	410	550	1022	800	1472	1080	1976	1413	2575		
218	425	552	1025	802	1475	1090	1994	1420	2588		
220	428	560	1040	808	1486	1093	2000	1427	2600		
230	446	566	1050	810	1490	1100	2012	1430	2606		
232	450	570	1058	816	1500	1107	2025	1440	2624		
240	464	580	1075	820	1508	1110	2030	1441	2625		
246	475	580	1076	830	1525	1120	2048	1450	2642		
250	482	590	1094	830	1526	1121	2050	1455	2650		
260	500	593	1100	840	1544	1130	2066	1460	2660		
270	518	600	1112	843	1550	1135	2075	1469	2675		
274	525	607	1125	850	1562	1140	2084	1470	2678		
280	536	610	1130	851	1564	1149	2100	1480	2696		
288	550	620	1148	857	1575	1150	2102	1482	2700		
290	554	621	1150	860	1580	1160	2120	1490	2714		

C.	F.
1° =	1.8°
2° =	3.6°
3° =	5.4°
4° =	7.2°
5° =	9.0°
6° =	10.8°
7° =	12.6°
8° =	14.4°
9° =	16.2°

Formulæ for converting from one scale into the other:

$$F = \frac{9C}{5} + 32;$$

$$C = \frac{5(F - 32)}{9}$$

F = degrees Fahrenheit.

C = degrees Centigrade.

Note to Table 45.—In order to comply with the title of this table, certain of the numbers have to be given twice, for an obvious reason. For instance 330° C. is the nearest whole number of degrees Centigrade corresponding to 625° F., but 626° F. is the nearest whole number of degrees Fahrenheit corresponding to 330° C.

APPENDIX III

§845. Notes to Text

Note A to pp. 57 and 59.—To calling carbonless and very low-carbon steel either "ingot iron" or "iron" there is the objection that each of these names is ambiguous, because it is capable of being used either specifically or collectively (§59). "Iron" has, besides its collective sense, two other specific senses already in wide use, "wrought iron" and "cast iron." Indeed the attempt to call the very low-carbon A.R.M. steel "iron" seems to have led to confusing it with wrought iron, or at least to inferring wrongly that it has some of the specific qualities which have given wrought iron its special fields of usefulness. All classes of iron and steel are "iron" in the collective sense. In such a work as this each word should be used as far as possible in a single sense, which in the case of "iron" must needs be the collective one. This requires us to call the various products by their specific names, the unambiguous one for this low-carbon molten-origin product being "steel."

Note B to p. 185.—That the exposure to the strong magnetic field does not, as it conceivably might, act through causing the allotropic iron of the hardened steel to revert to the alpha and magnetic state, in short by drawing the temper, I showed by having a piece of steel of 0.92 per cent. of carbon, which had been hardened by quenching in cold water from 830°, next exposed to a magnetic field of 11,500 C.G.S. through the kindness of Professor W. I. Slichter. The Shore hardness before this magnetization varied, according to 15 determinations, only from 56 to 57. Within 10 minutes after the magnetization the Shore hardness was 56, and at a few spots near the center of the piece 54, or practically identical with the initial.

Note C to p. 186.—Because the specific magnetism of the quenched high-carbon steel of Hadfield and Hopkinson lacks only about 11 per cent. of equalling that of soft iron, the hard element which gives that steel its great hardness must be magnetizable in strong fields, because so little as 11 per cent. of an unmagnetizable substance can hardly be supposed to give such great hardness. Because the specific magnetism in like fields of their manganese steel, hardened to a ball hardness of 418 by a 650-hour exposure to 650, is only 0.2 per cent. of that of soft iron, the hard element that gives this steel its hardness cannot be magnetizable in these same fields, and hence differs in this respect from the hard element of quenched carbon steel. Hence the need of postulating two hard or beta forms of iron, one unmagnetizable in any field, beta I, and one magnetizable in strong but unmagnetizable, or much less magnetizable, in weak fields, differing from alpha iron in this respect, in its hardness, and apparently in its silhouettes.

Our humanly natural reluctance to admit the complexities of nature, and in particular to admit the polymorphism of iron, should not take the form of moral cowardice like that which leads us to hide the truth from our physician. It should weaken before the disclosures of Cohen, who shows us that our common metals, such as antimony, bismuth, copper, lead, and zinc, as we buy and sell them, are complexes of two or more allotropic forms. If this is true of these metals which at first seem patterns of regularity, what may we expect of the erratic iron? ("The Influence of the Metastability of Metals," *Trans. Faraday Soc.*, 1915, vol. 10, p. 216.) Many proceed by postulating certain properties which they imagine that allotropic iron ought to have, and then denying its existence because they do not find these properties. The better course is to infer the properties from the phenomena.

Note D to p. 189.—The wide variations possible in the magnetism which accompanies a given degree of hardness is shown by comparing the data of Hadfield and Hopkinson, Figs. 31, V and VI with those of Gumlich ("Magnetic Properties of Iron-carbon and Iron-silicon Alloys," *Trans. Faraday Soc.*, 1912, vol. 8, pp. 1 and 104). A Brinell hardness of 418 is accompanied by a specific magnetism of 0.2 per cent. in the former, but of 95 per cent. in the latter set of data or about 475 times as much. Gumlich gives 95 per cent. by interpolation as the specific magnetism of steel of about 0.36 per cent. of carbon, which is the carbon content to which a hardness of 418 in the quenched state corresponds.

Note E to p. 211.—Many and perhaps all series of liquids have, in addition to the liquidi such as AB and BD of Fig. 23, p. 130, at which solidification starts in the presence of nuclei of the solid, lines, $A'B'$, $B'D'$, Fig. 124, roughly parallel to them, which may be called supersolubility liquidi, at which solidification starts even in the absence of nuclei, at least provided that there is agitation to start it. Above ABD the liquid is stable; in the space between ABD and $A'B'D'$ it is metastable, with its tendency to solidify restrained by lag; and if it cools to below these last it is labile, with that tendency needing only agitation to become self executing. The intersection of $A'B'$ and $B'D'$ is called the hypertectic point. A liquid, say R, might cool past the liquidus BD because of the

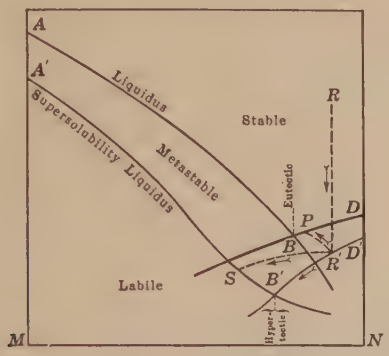


FIG. 124.—Possible paths of undercooling solidification. See Miers and Isaac.[1]

lack of nuclei to start solidification. On reaching the supersolubility liquidus at R' it would, at least if there were agitation, begin to solidify by the precipitation of substance N. According to the rapidity with which heat is now evolved by this precipitation the liquid may follow various paths, such as are indicated by the three arrows. It may complete itself at B, yielding the usual eutectic, or at B' yielding the hypertectic, which may differ in composition from the eutectic.

[1] See Miers and Isaac, *Proc. Royal Soc.*, 1907, A, vol. 79, p. 322; Isaac, *idem*, 1910, A, vol. 84, p. 344; and Desch, "Metallography," Longmans, London, 1910, p. 200.

SUBJECT INDEX

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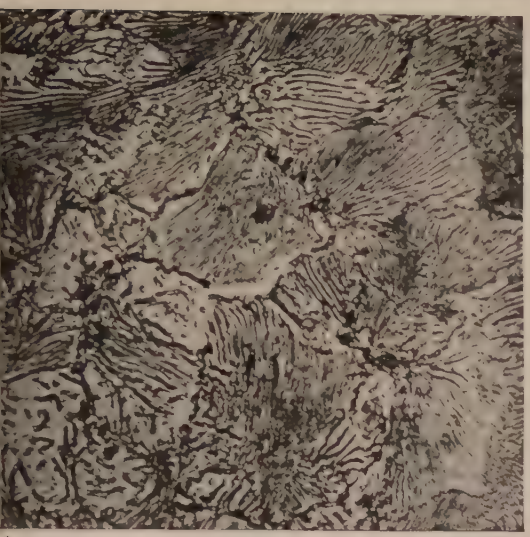
Chief Constituents of the Steel-White-Cast-Iron Series.



A. × 500

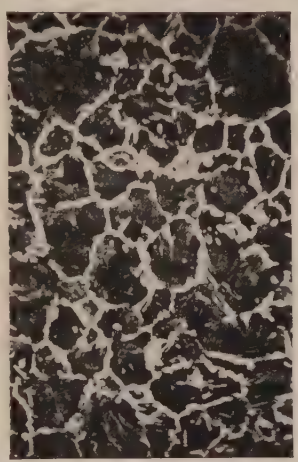
Pearlite. Carbon 0.92. §§71, 74.

B. × 500

Hypo-eutectoid steel. Carbon 0.59. White = ferrite shells, dark = pearlite kernels. §§71, 74.

C. × 500

Hyper-eutectoid steel. Carbon 1.20. White = cementite shells, dark = pearlite kernels. §§71, 74.



D. × 42

Hypo-eutectoid

E. × 47

Hyper-eutectoid

Steel. White pro-eutectoid shells, pearlite kernels. §75.

F. × 180

Eutectic in spiegeleisen. §81.

G. × 30

Hyper-eutectic cast-iron. White bands *a*, *b*, *c* of primary cementite; honeycomb *d* of eutectic. §§81, 175.

For description of plates see Table 38, pages 608 to 615. The authors of micrographs previously unpublished are credited thus: STEAD. Those of micrographs previously published are credited thus: *Rosenhain*.

PLATE 2

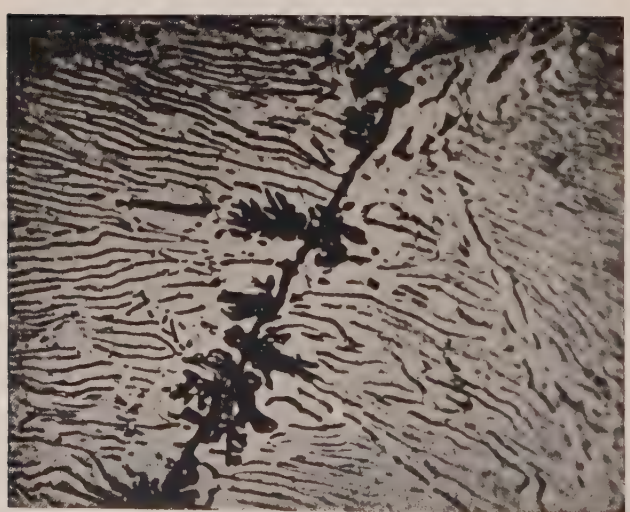
Constituents Second In Importance.

A. × 100
Martensite (black) in austenite.—STEAD. §99.

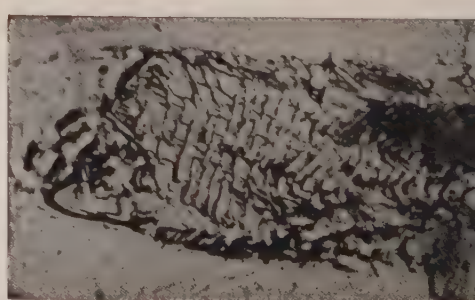
B. × 500
Ferrite F, Martensite M, Troostite black. §97.

C. × 3300
Hyper-eutectoid steel. Carbon 1.14. Etched with sodium picrate. §§71, 74.
Black = cementite, white = pearlitic ferrite.

E. × ¾ (about)
Primaustenoid crystal.—TSCHERNOFF. §187.



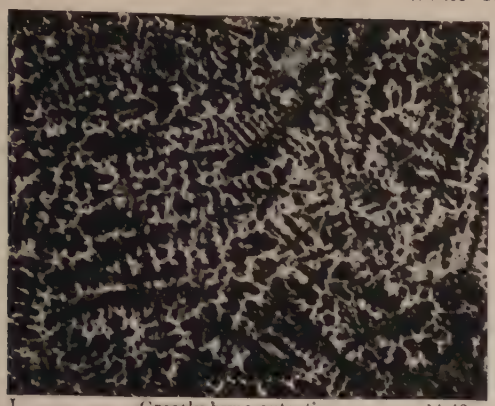
D. Same as C. × 1200



F. Crumpled pearlite. §108. × 1200

A, B, and F, etched with picric acid; C and D with sodium picrate.

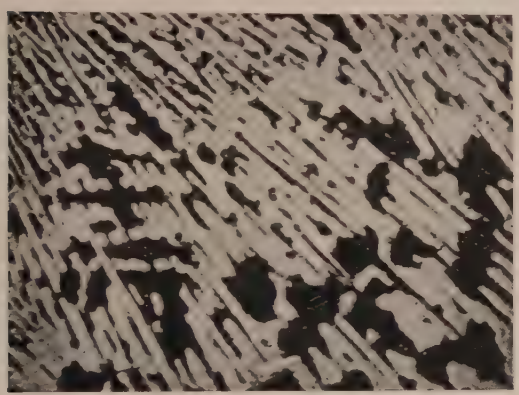
White Cast-Iron, §§92, 196, 205.



J. Greatly hypo-eutectic. × 42

K. Moderately hypo-eutectic. × 42

White = eutectic, dark = primaustenoid.



L. Slightly hypo-eutectic. × 180

M. × 180

Eutectic. White = eutectic cementite, dark = eutectic austenoid.

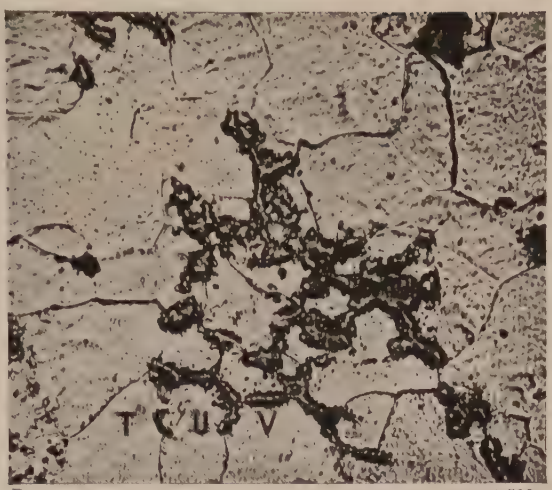
N. Spiegeleisen. × 17

O. Quenched in water.—*Tiemann.* × 195

Hyper-eutectic. White blades = primary cementite; honeycomb = eutectic.

Gray Cast-Iron = A Steel or White Cast-Iron Matrix + Graphite

A. × 580



B. × 500

Black heart malleable casting, light = ferrite, dark = graphitiferous area. Matrix equivalent to very low carbon steel.—W. Campbell. §§118, 123.

C. × 90

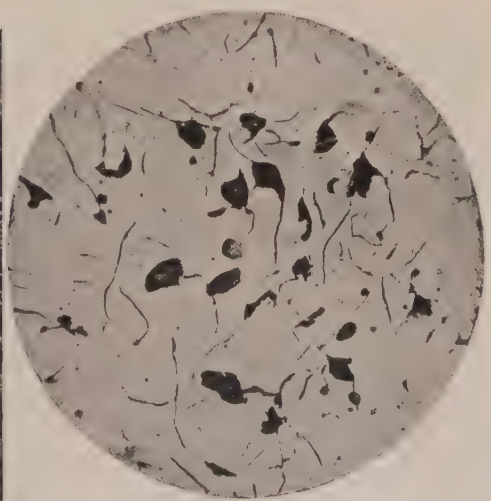
Black heart malleable.—W. Campbell. §§118, 123.

D. × 500

Gray cast-iron with matrix equivalent to hyper-eutectoid steel. §120.

E. × 500

Gray cast-iron. Matrix equivalent at right to hypo-eutectoid steel, at left to white cast-iron. §118.



F. × 100

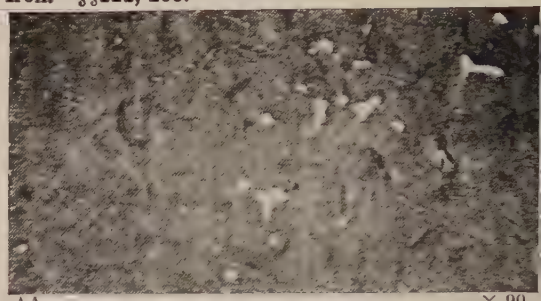
Compact graphite in Johnson's oxygenated cast-iron.—Waterhouse. §§124, 299.

PLATE 5

A,AA. Graphite in Cold-blast Charcoal Pig-Iron. B to E. Transformation Structures in One and the Same Hypo-Eutectic Cast-Iron. §§121, 295.

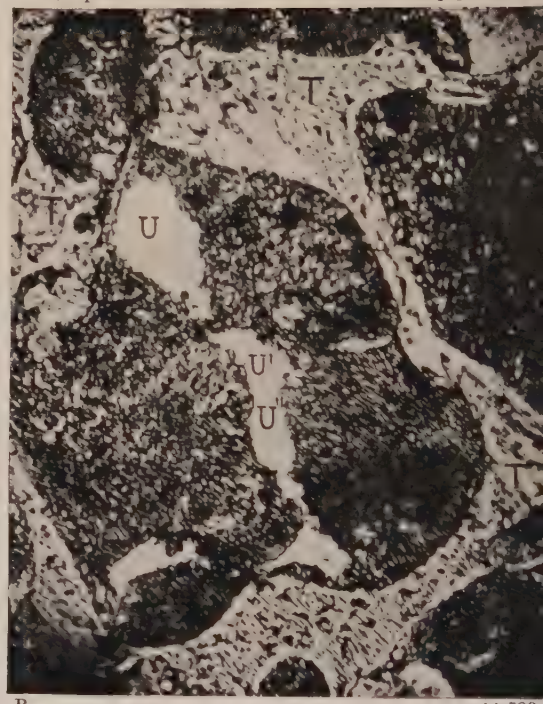
A. × 100

Graphite in cold blast charcoal pig-iron. §301.



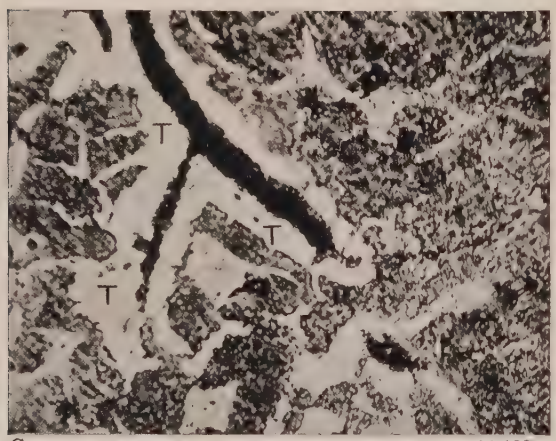
AA. × 90

Graphite in a low-carbon cold-blast charcoal pig-iron.



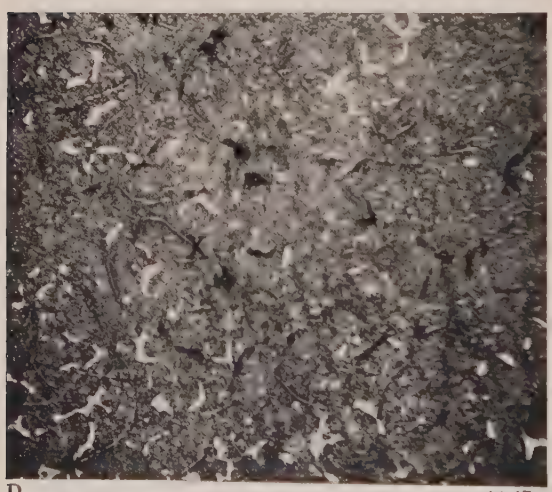
B. × 500

Network structure in white cast-iron. Carbon 3.055%. §§121, 295.



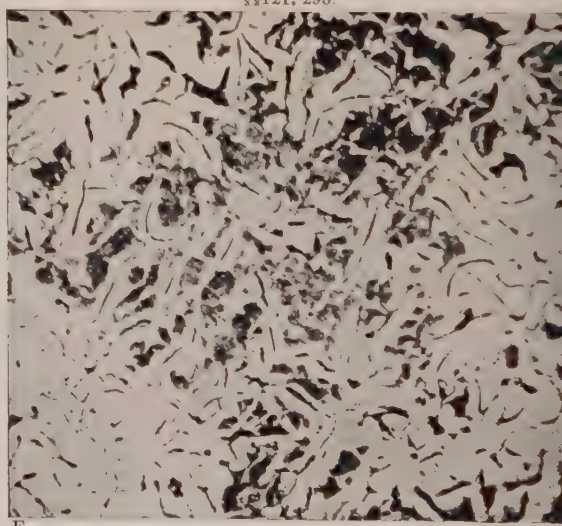
C. × 180

Eutectiform graphite and ferrite. §§121, 295.



D. × 47

Air-cooling under-accents graphitization.



E. × 47

Furnace cooling exaggerates graphitization.

Influence of rate of cooling on degree and prominence of graphitization. §§121, 295.

Austenite, Rock Eutectic and Graphite.

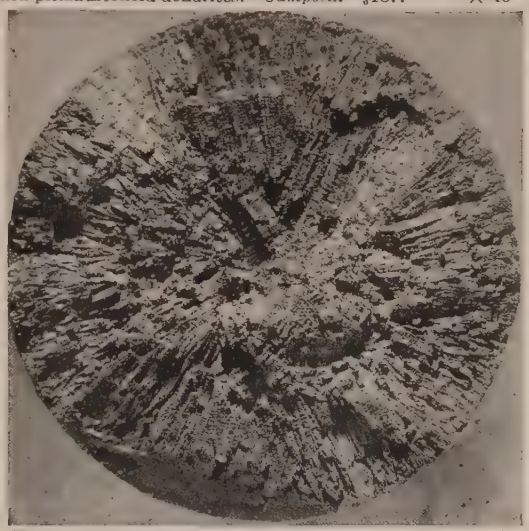
A. × 18

Primaustenoid dendrites, sodium picrate. §187.

B Black primaustenoid dendrites.—*Campbell.* §187. × 40

× 142

Primaustenoid dendrites.—*Hadfield and Hopkinson;* picric acid §187.



D. × 1

Primary austenite dendrites in hot fracture. §184.

Eutectic of quartz and orthoclase, surrounding primary crystals of acidic plagioclase.—KEMP. 106.

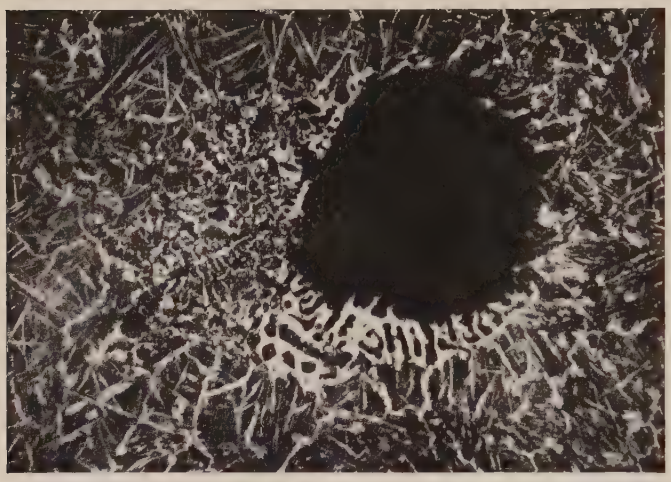
F. Deep-seated. ×47

G. Superficial. × 47

Graphite content and coarseness inverse to rate of cooling. §101, 302.

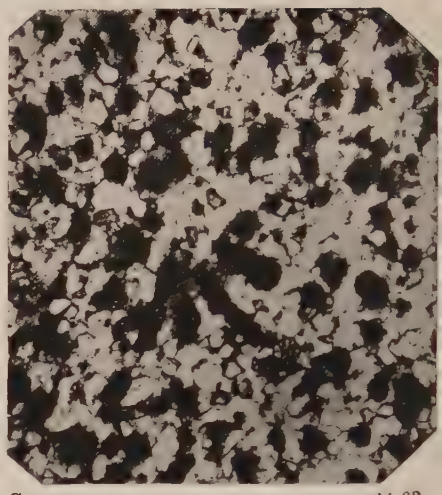
PLATE 7

Graphite (Figs. C, D, F, G), and Undue Eutectic (Figs. A, B, E).



A. × 60

Eutectic rim in 1.08 per cent. carbon steel. §209.



C. × 63

Typical black heart malleable.—Touceda. §118.

B. × 800

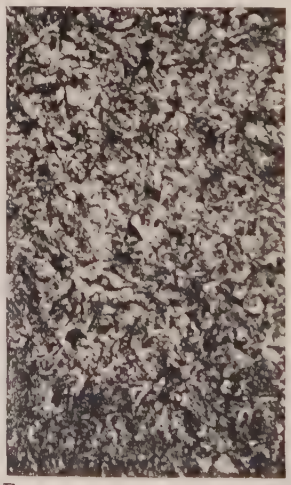
Eutectic from steel of 1.08 per cent. carbon. Black primaustenoid dendrites in zebra eutectic matrix. §209.

D. × 42

Ferrite rings about graphite centers in 1.45 carbon steel. §289.

E. × 42

Eutectic in steel of about 1.60 carbon. §209.



F. × 42

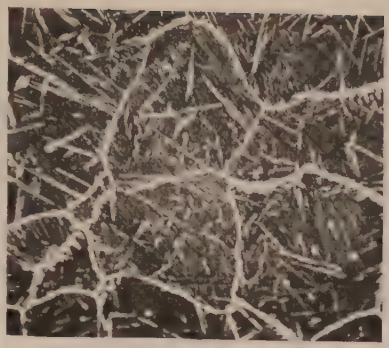
Fine-grained gun-metal. §297.

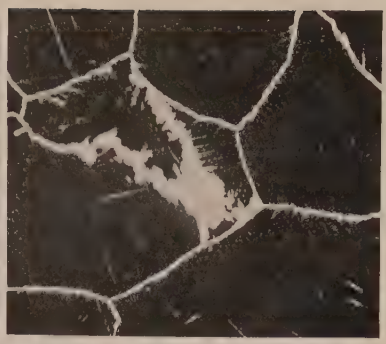
G. × 42

Ferrite rims about graphite ghost lines in 1.45 carbon steel. §§108, 289.

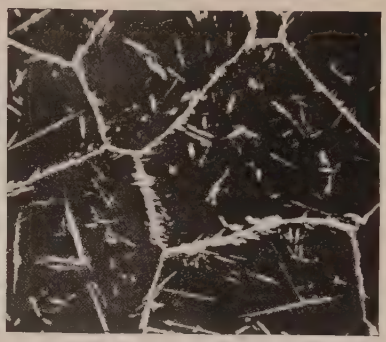
po-eutectoid Steel Shown Micrographically.

D. 790°. 1 hr. 15 min.





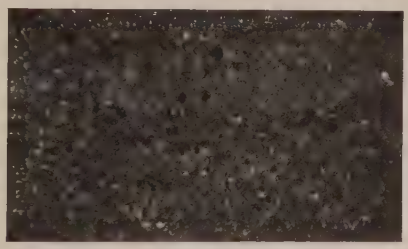
F. 740°. 1 hr. 45 min.



H. 705°. 2 hrs. 15 min.

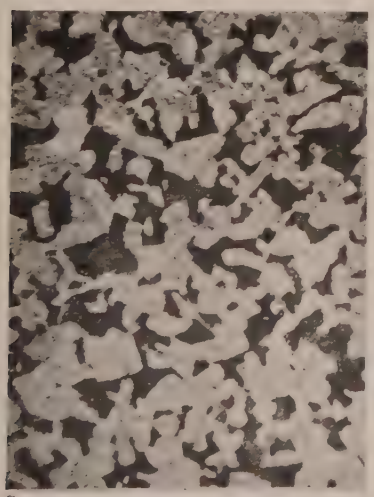
Dark = pearlite.

M. 925°.



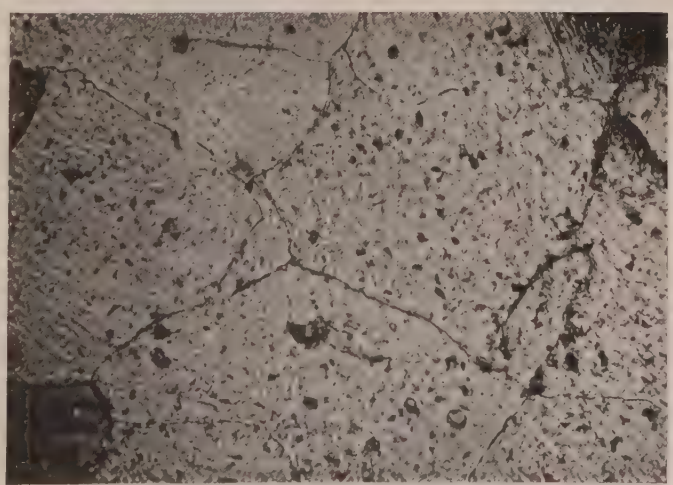
N. 950°.

censite.



S. × 42

705°. 2 hr. 28 min.



T. × 500

705°. 2 hr. 28 min. Ferrite grains.

Dark = pearlite.

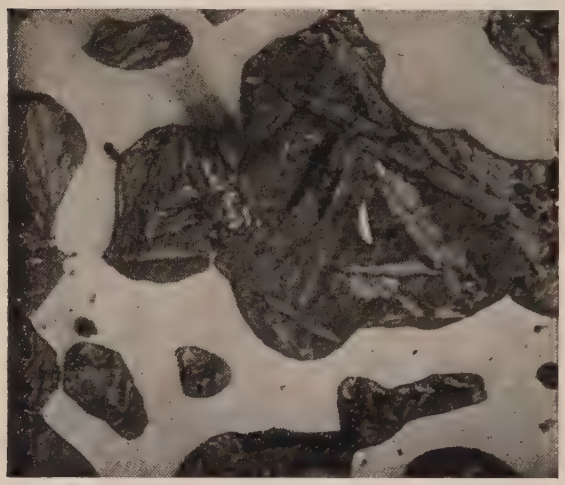
Incipient Fusion in Steel of 1.60 per cent. Carbon.—*Zimmerschied.* × 500

A. Quenched from slightly above solidus. Much twinning. §264.

B. Quenched from a temperature higher than that of Fig. A. Martensite grains, eutectic between grains. §265.

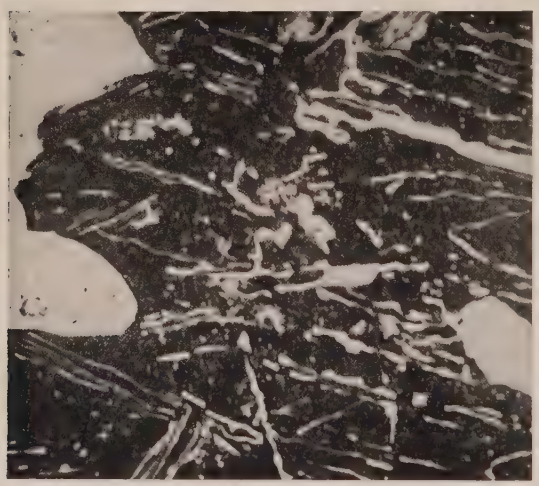
PLATE 11

The Transformation in White Cast-Iron. Precipitation of Pro-eutectoid Cementite in Cooling. Quenched from the temperature indicated. §272. × 1200.

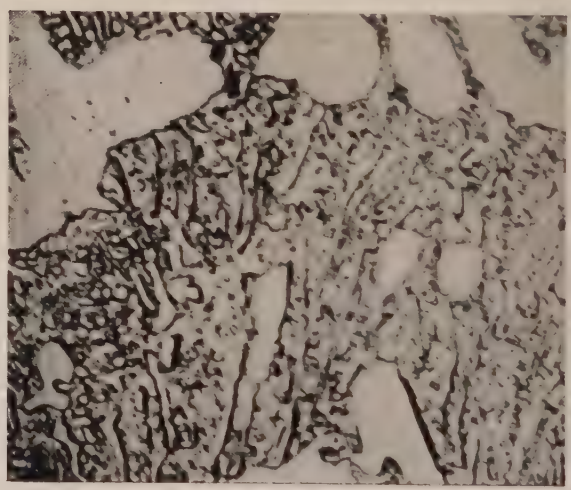


A. 1050°. White = eutectic cementite, black = martensite.

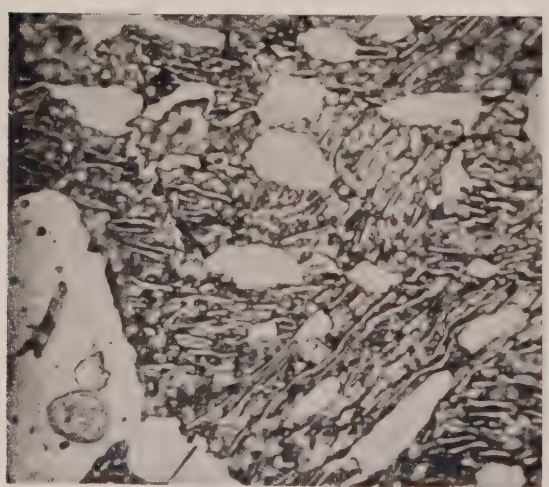
B. 765°. Cementite + martensite + pro-eutectoid cementite needles.



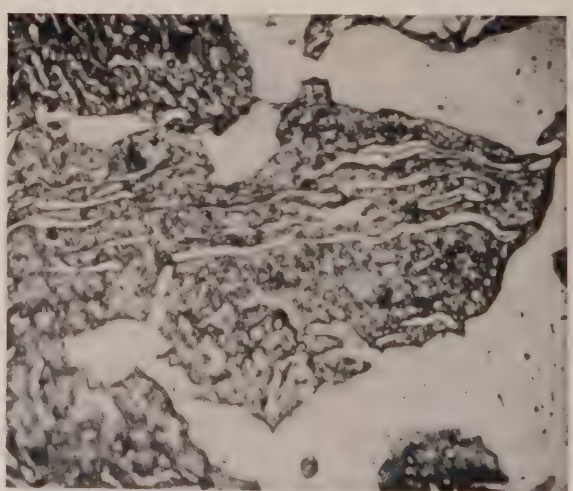
C. 728°. Cementite + troostite + cementite needles.



D. 707°. Cementite + divorced pearlite + cementite needles.



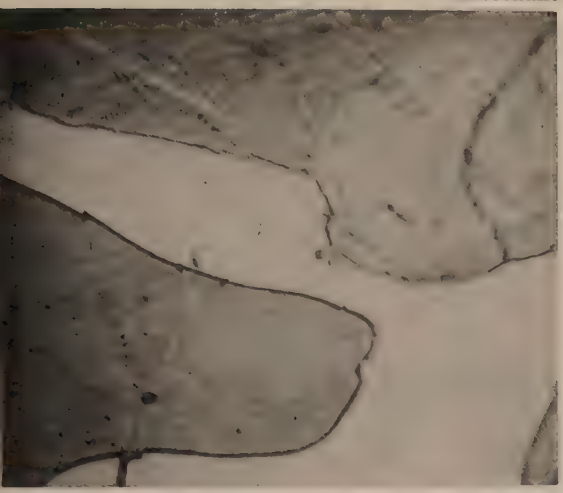
E. 671°. Same as Fig. D.



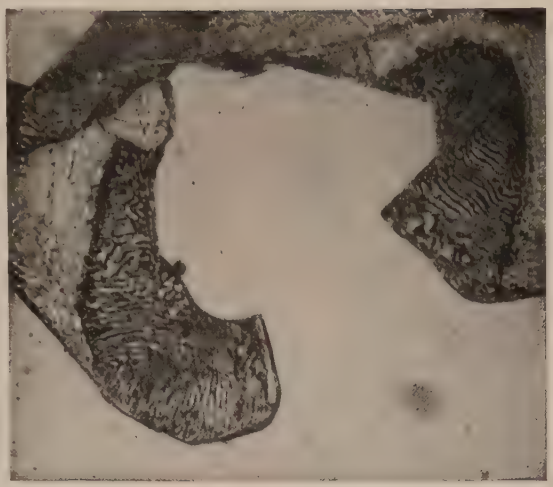
F. 646°. Same as Fig. D.

Most of the eutectic is here replaced by eutectiform cementite, probably because of under-cooling.

Miscellaneous Cast-Iron.



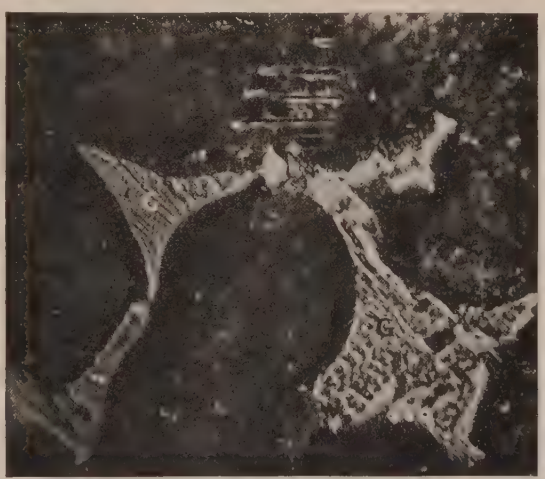
A. Quenched from 1050°. White = eutectic cementite, dark = eutectic austenite.



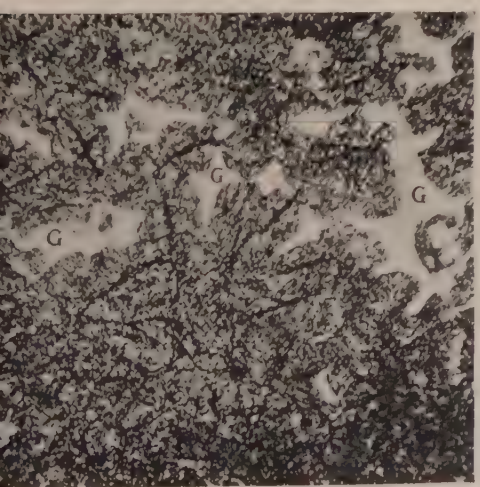
B. Cooled slowly. White = eutectic cementite, dark = eutectic-austenoidal pearlite.

The transformation in the eutectic austenite. §269. × 1200.

C. × 500
Pro-eutectoid cementite in primaustenoid. §272.



D. × 500
Typical eutectic outlines. §274.



E. × 180
Thin casting (radiator iron).

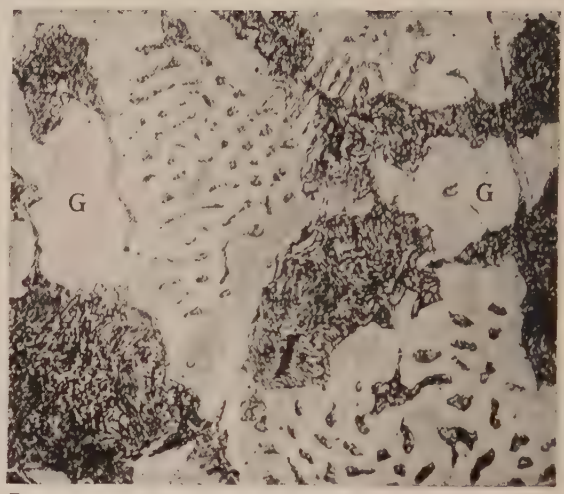
F. × 180
Thicker casting (car wheel).

Variations in coarseness of graphite in industrial castings. §§125, 289, 306.

PLATE 13

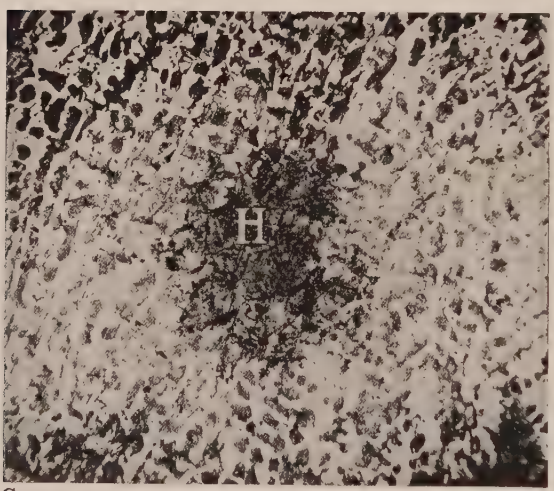
Graphitization Increases with Slowness of Cooling. Unusually Good Chilled Cast-Iron Car Wheel Cast by A. W. Whitney.

A. × 47

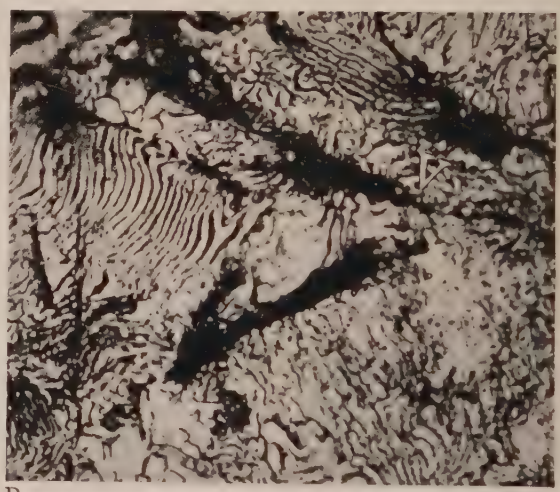


B. × 500

White cast-iron, the chilled face of tread of wheel. §304.

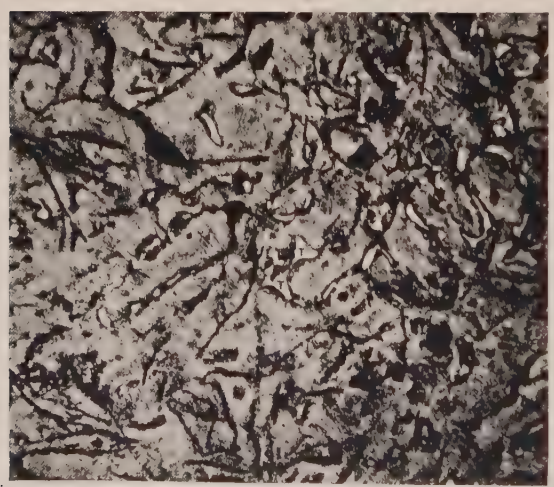


C. × 47



D. × 1200

Mottled cast-iron, one inch back from face of tread. §§123, 304.



E. × 47

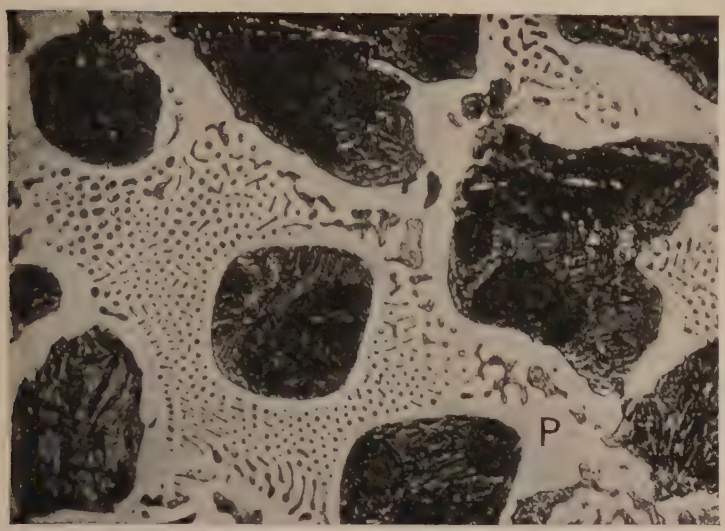
Graphite, pearlite, and eutectic.

F. × 500

Eutectic, *T*, pearlite, *S*, and slag, *V*.

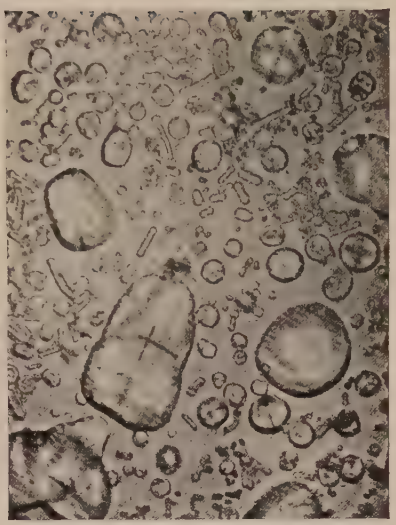
E and F are gray cast-iron from interior or "plate" of wheel. §§274, 304.

Cast-Iron and Miscellaneous.



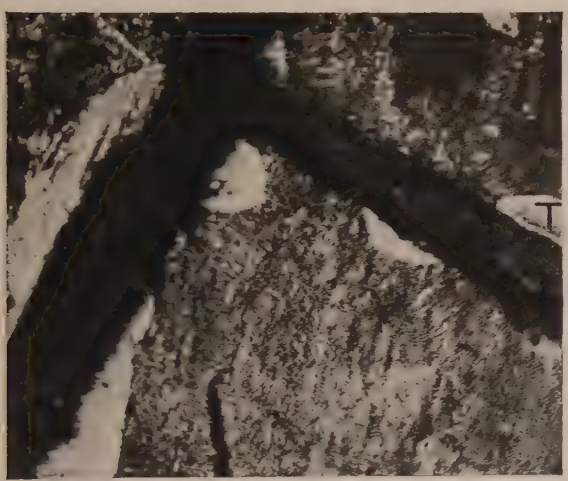
A. × 500

Undercooling cementite in chilled cast-iron. §274



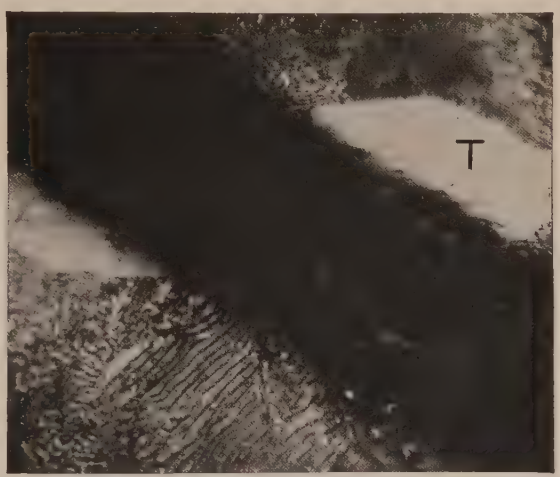
B.

Liquid crystals.—*Lehmann.* §§333, 704.



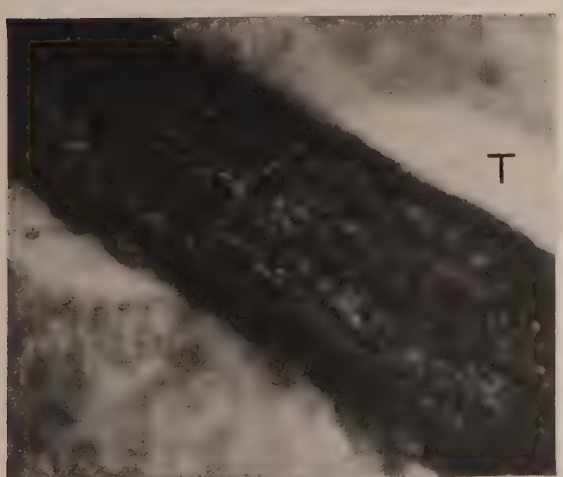
C. × 500

Bad coke carwheel cast-iron. §§123, 125.



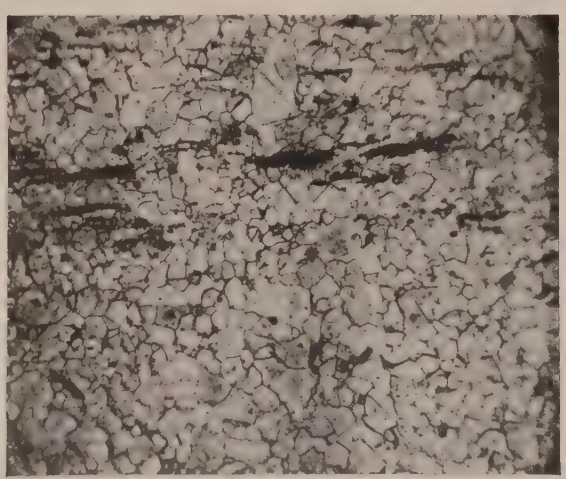
D. × 1200

Same field as C, focussed on pearlite. §§123, 125.



E. × 1200

Same as D, focussed on skeleton of ferrite in black graphite band. §§123, 125.

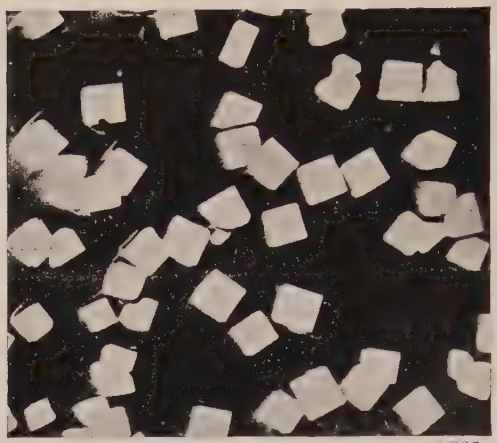


F. × 50

Grain structure and slag in wrought-iron. §§50, 766.

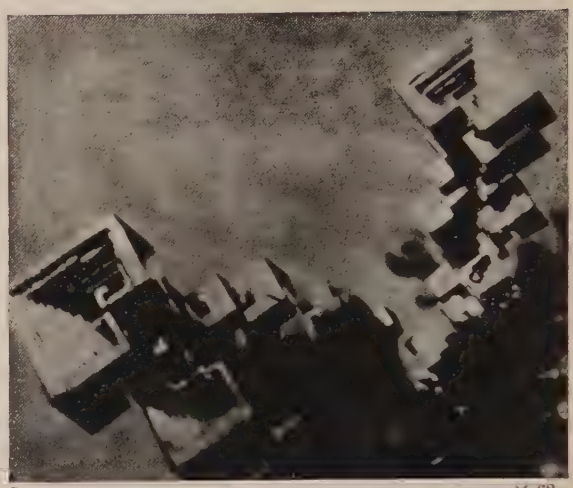
PLATE 15

Idiomorphic and Dendritic Crystals, and Cubic Cleavage in Iron.



A. × 235

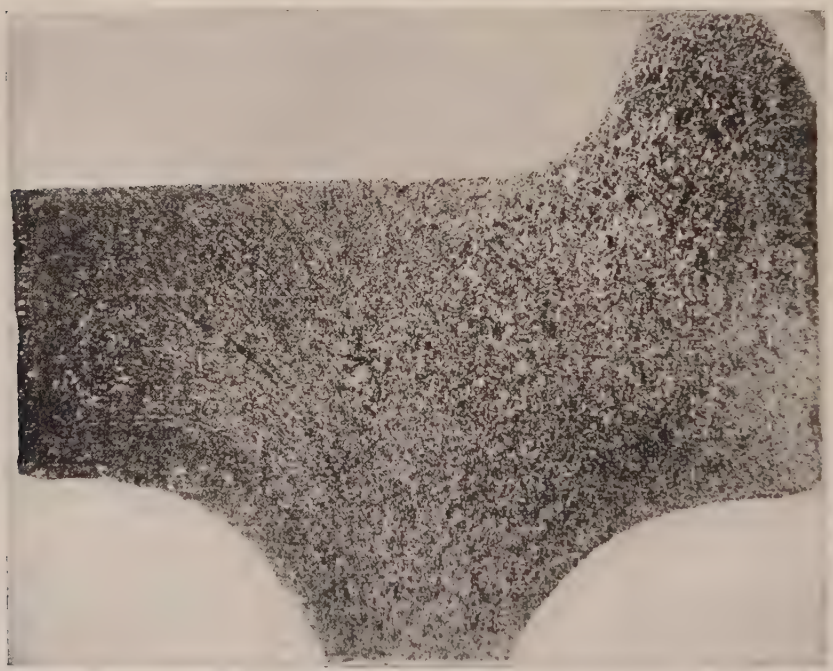
Cubes of ferrite. §342.



B. × 62

Dendritic cubes of austenite. §342.

Osmond and Cartaud.

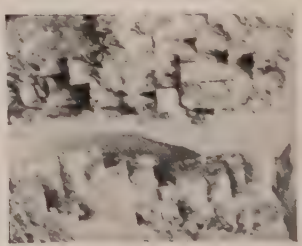


C. × 1

Dendritic structure in rolled carwheel. §§343, 762, 796.

D. × 150

Cementite crystals in cast-iron. —*Stead.* §342.

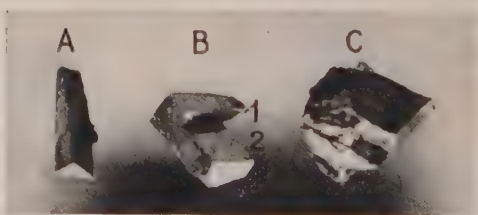


F. × 2.5

Cubic fracture, silicon steel. §§365, 738.

E. × 90

Augite dendrite.—*Rosenbusch.* §343.



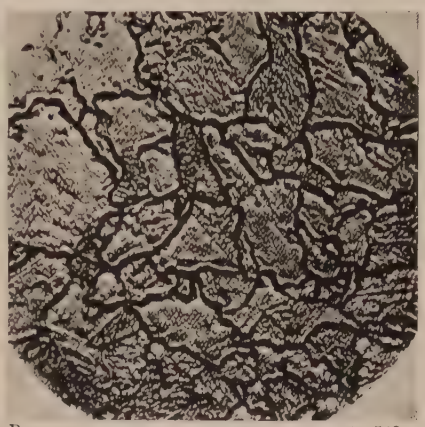
G. × 1.3

Cubic fracture in ferrite. §§365, 449, 738.

Crystalline Structure (A,B,C,E) and Expulsion of Pro-eutectoid Cementite (D,F).

A. × 1

Coarse-grained silicon steel. §345.



B. × 540

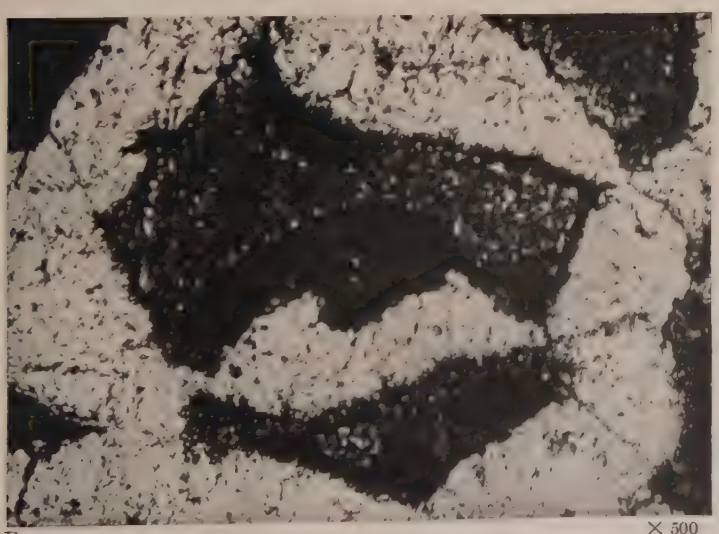
Austenite etched at 1050°.—*Wark.* §§345, 361, 371.

C. × 15

Dendrites in antimony.—CAMPBELL. §344.

D. × 500

Expulsion of Fe_3C masks pre-existing slip-bands. §368.



E. × 500

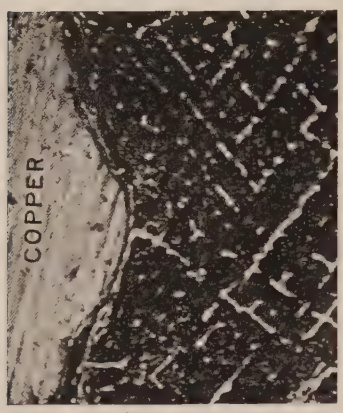
Troostite, black; martensite, gray; ferrite, white. Ferrite grains in the network. §371.

F. × 50

Pro-eutectoid cementite assembled in grain boundaries and octahedral planes. Manganese steel. §§368, 405.

PLATE 17

Expulsion of the Pro-eutectoid Element (A,B,C,D,F). Etching Figures in Ferrite (E,G,H).



A. × 500

Fe_3C expelled to outer surface, then copper plated.

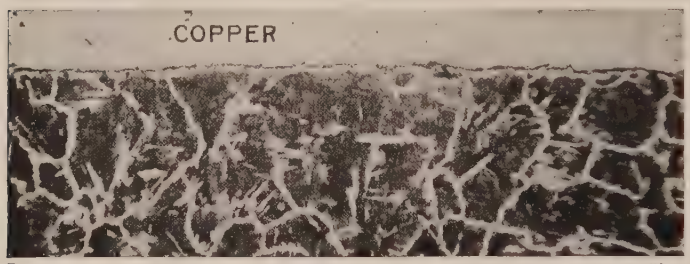
B. × 50

Fe_3C expelled to internal surfaces.

Manganese steel. §368.

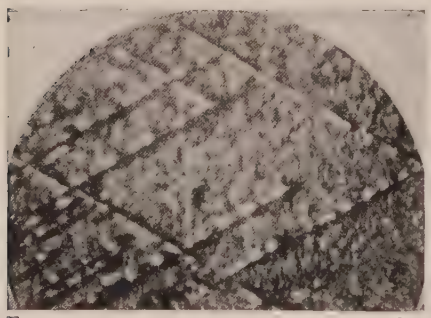
C. × 50

Ferrite expelled after slag to cavities between idiomorphic crystals in ingot pipe. §368.



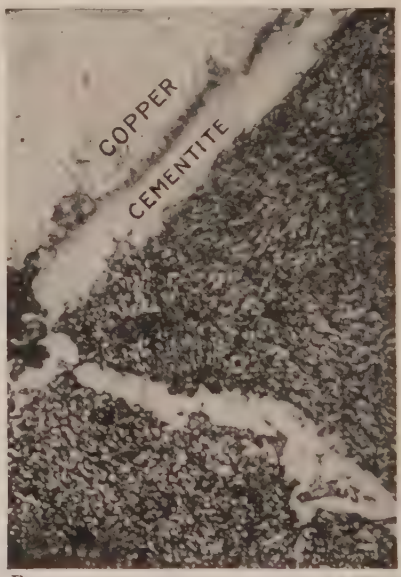
D. × 50

Fe_3C expelled to outer surface of steel of 1.45 % carbon, then copper plated. §368.



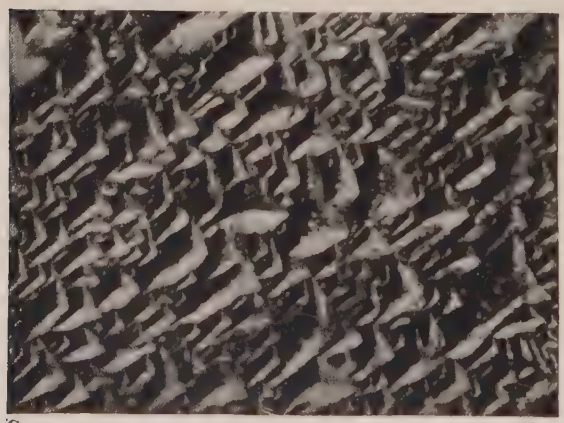
E. × 100

Figures from deeply etching ferrite.—*Stead.* §§371, 374.



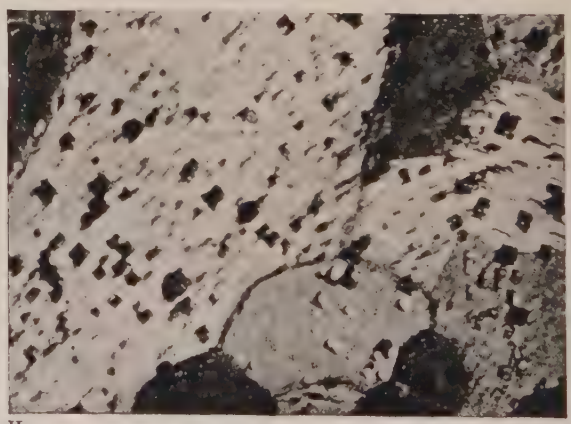
F. × 500

Same as D. §368.



G. × 130

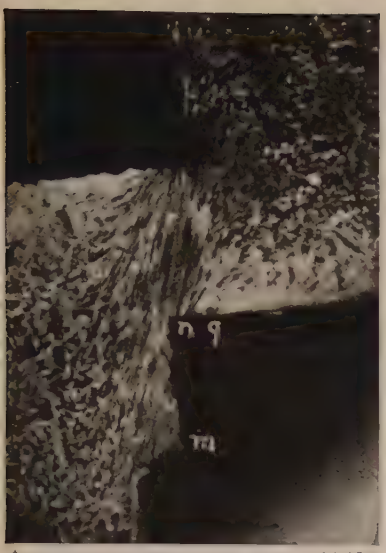
Rectangular etching cavities in silicon steel.—*Stead.* §§371, 374.



H. × 60

Cubic etching pits in silicon steel.—*Stead.* §371.

Etching and Pressure Pits in Ferrite.

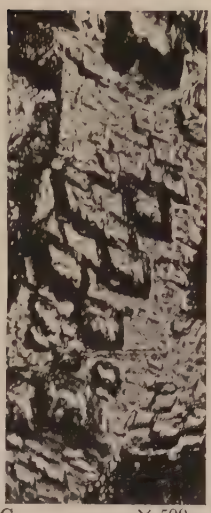


A. × 18

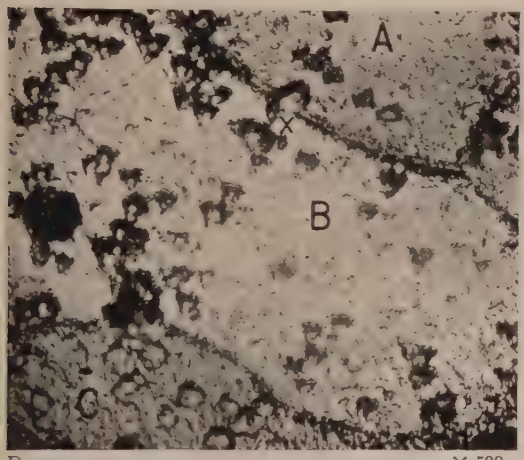
Ferrite hairs caused in punching.—*Howe and Levy.* §§108, 373, 532.

B. × 500

Geometrical etching pits in ferrite hairs of Fig. A. §§371, 373.

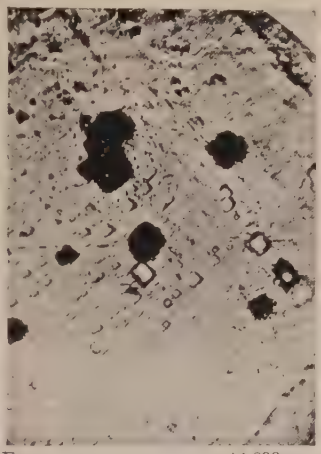


C. × 500



D. × 500

Shape of etching pits changes from grain to grain. §§371, 709.

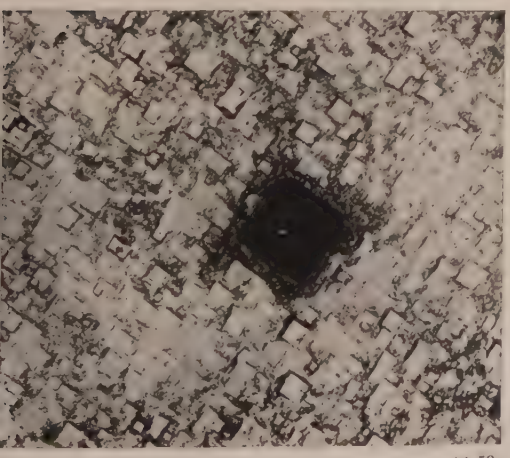


E. × 800 ±

Cubic etching pits in cold-rolled iron.—*Ewing and Rosenhain.* §§371, 373.

F.

Pressure squares round holes.—Stead. §379.



G. × 50

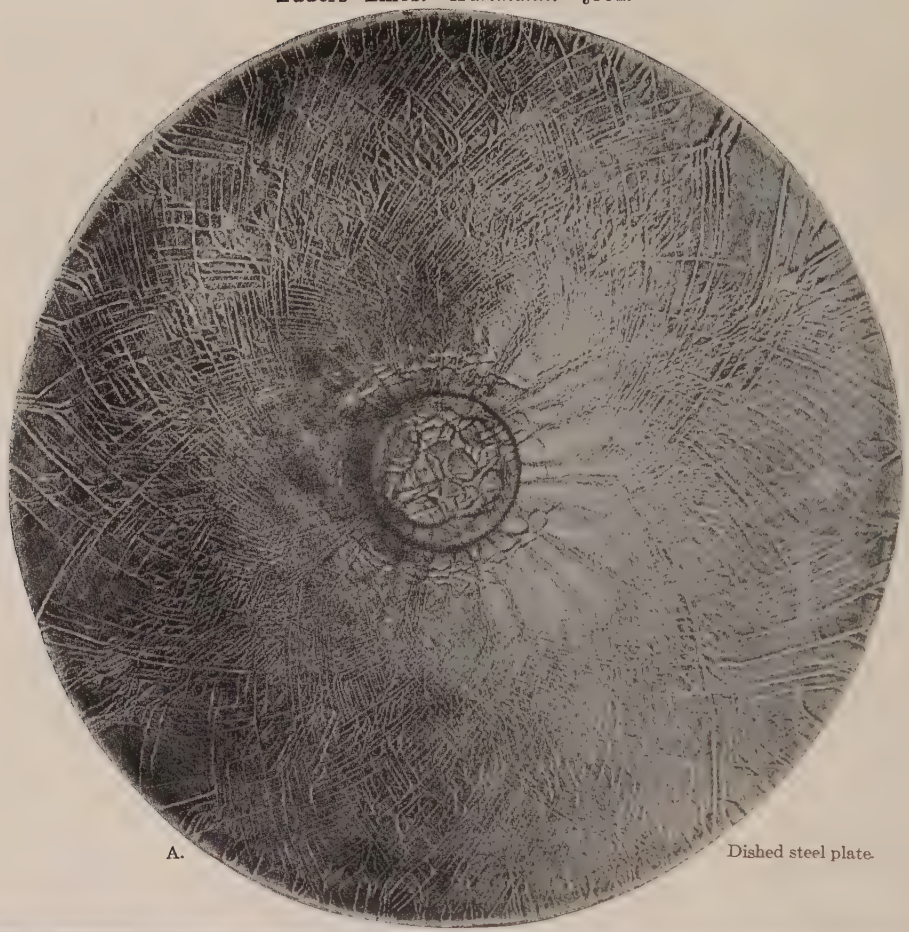
Etching squares a round hole.—Stead. §371.

H. × ½

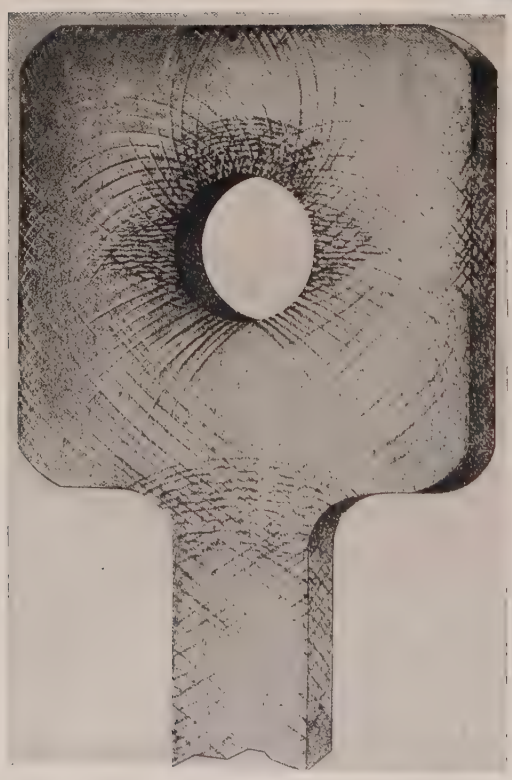
Abrupt change from columnar to equi-axed grains. §776.

PLATE 19

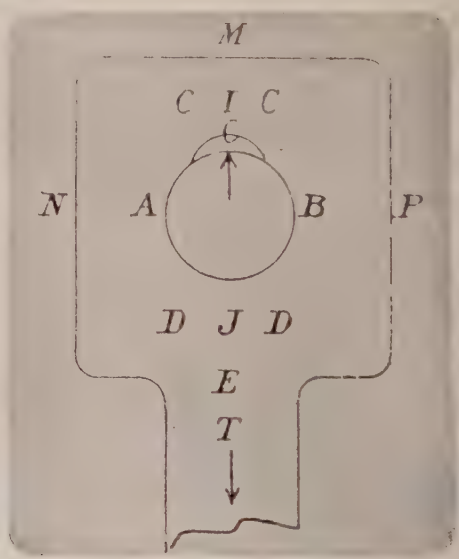
Lüders' Lines.—*Hartmann.* §391.



A. Dished steel plate.



B. Tensile test bar.



C. Key to B.

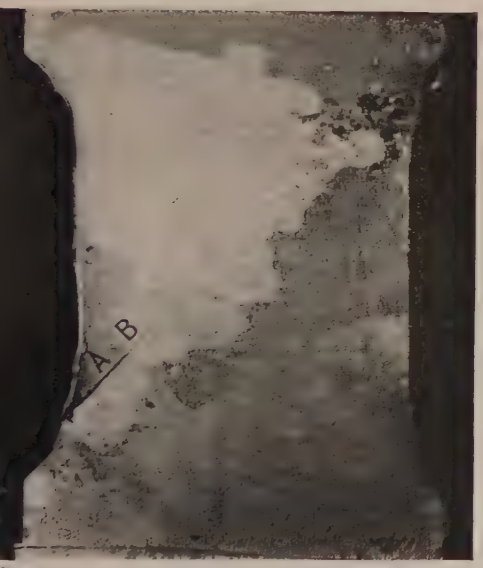
Lüders' Lines.

A. × 1

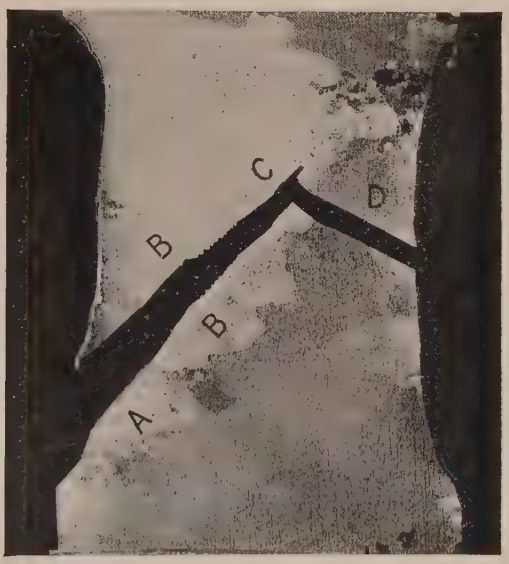
Radial and circular lines in glass. §391.

B. × 29.

Sawteeth near B of Fig. D.

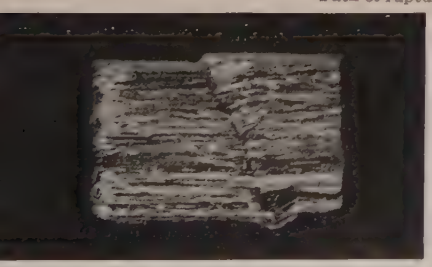


C. × 1.1



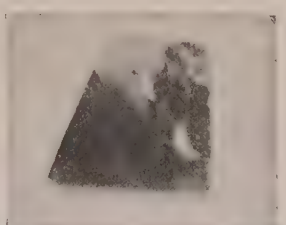
D. × 1.1

Path of rupture in Ruder's coarsened silicon steel. §§396, 399.



E. × 2.7

Lüder's lines in compressed wood. §395.



F. × 1.5

Wedge test for progress of growth of slip-bands. §397.

PLATE 21

Grain Uplift in Ferrite and Austenite.

A. × 50

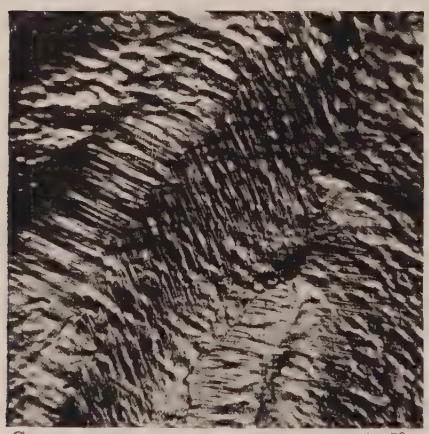
Accordion markings on polished surface. §410.

AA. × 6

Persistence of columnar structure through 1100° quenching in Mn steel. §408.

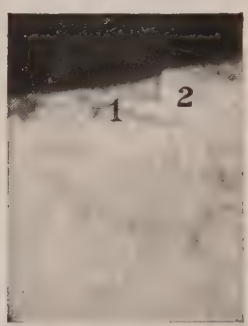
B. × 50

Surface crumplings in manganese steel. §411.



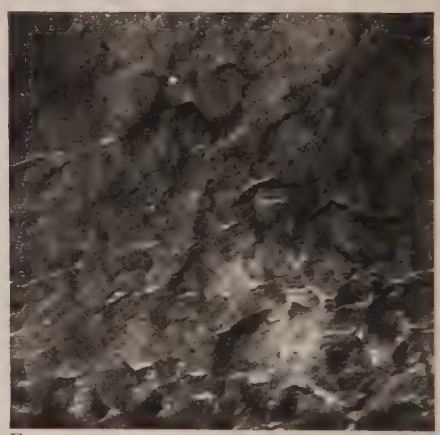
C. × 50

Slip-bands change direction at boundaries of columnar grains in manganese steel. §§405, 408.



D. × 1200

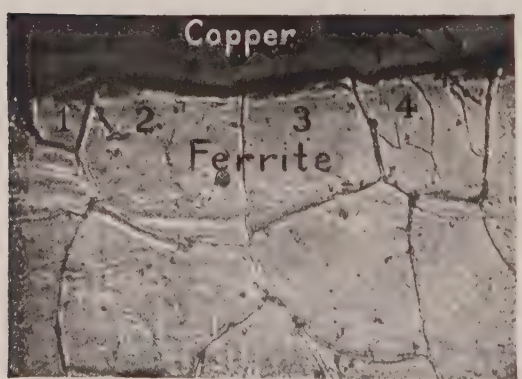
Cross-section.



E. × 50

Plan.

Grain-uplift in ferrite. §407.

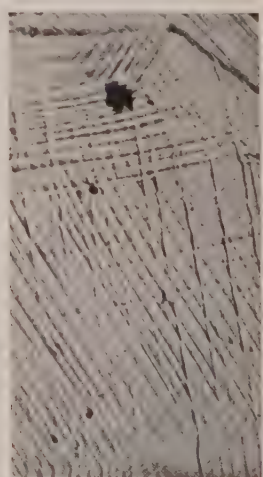


F. × 500

Cross-section.

G. × 500

Plan.

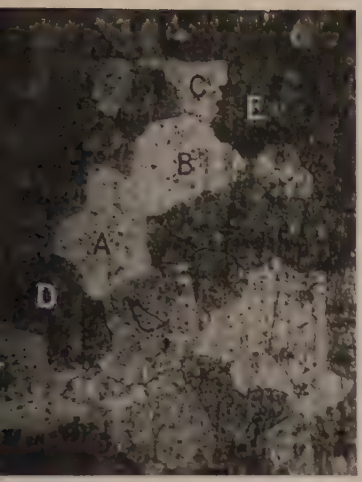


H. × 50

Four directions of slip-bands.—Invar. §417.

Grain-uplift in ferrite. §407.

Slip-bands in Ferrite and Miscellaneous.



A. × 42
Vertical light.

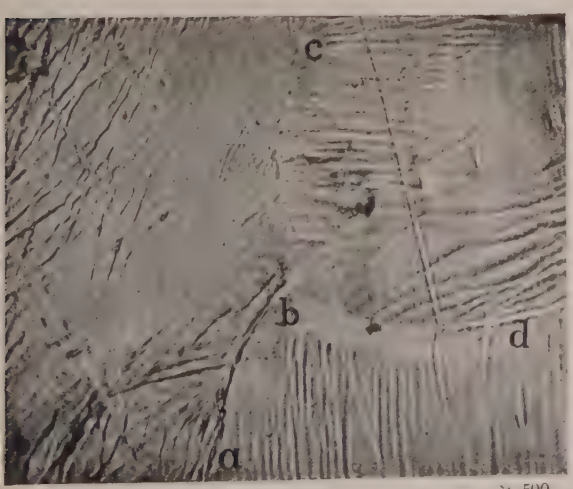
B. × 42
Oblique light. Same field as A.

C. × 500
Irregular slip-bands in ferrite. §412.

Relative brightness of ferrite grains changes with the light. §380.

D. × 500
Regular slip-bands in ferrite. §§412, 499.

E. × 500
Irregular slip-bands in ferrite. §412.



F. × 500
Slip-bands in ferrite. §§408, 412.

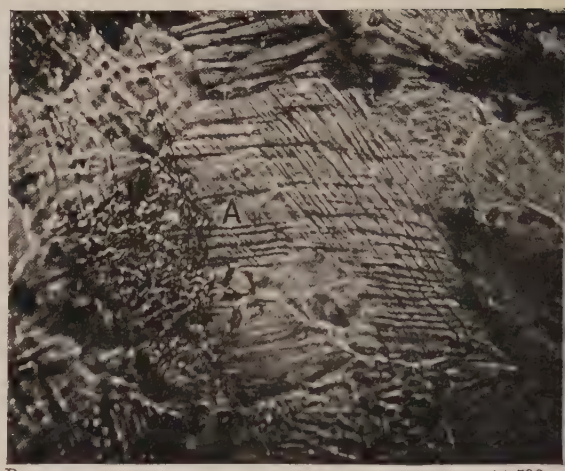
G. × 50
Rounded projections in austenite. §410.

PLATE 23

Slip-bands in Ferrite (All Etched Except B). §§412, 497, 503.

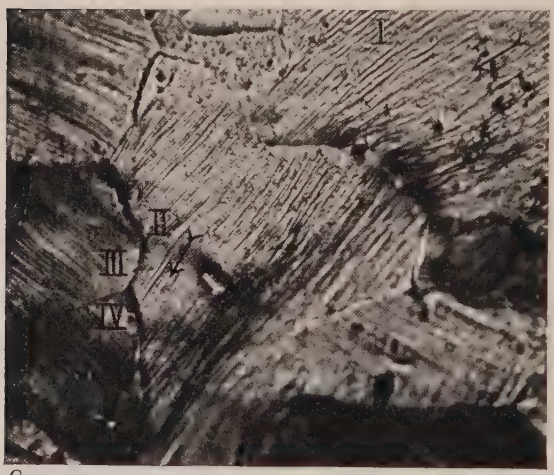
A. × 500

Slip-bands in etched ferrite.



B. × 500

Slip-bands fault each other.

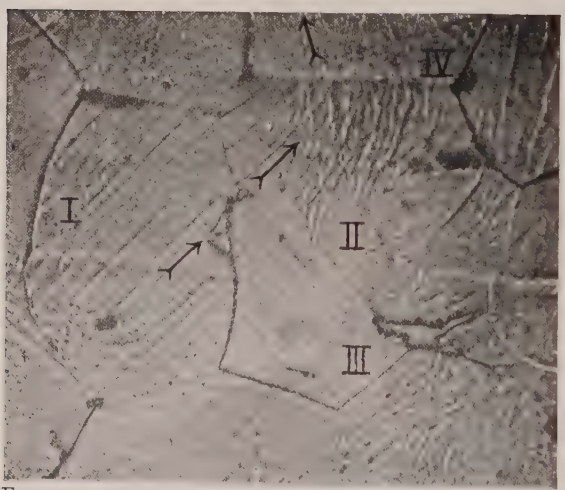


C. × 500

Slip-bands in etched ferrite.

D × 500

Slip-bands cross grain boundaries in etched ferrite.



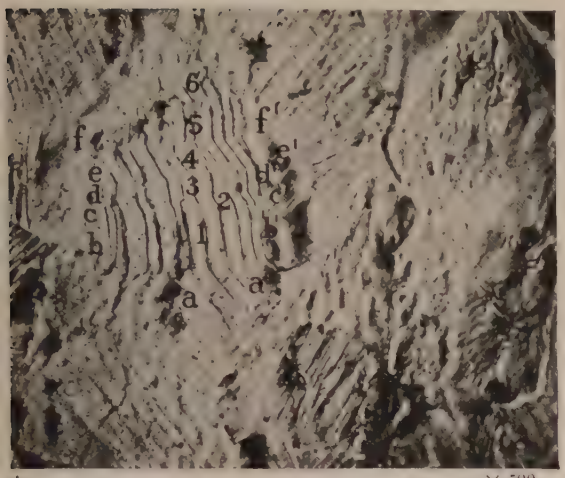
E. × 500

Slip-bands curl to cross grain boundaries.

F. × 500

Slip-bands not reaching terminal boundaries.

Slip-bands and Grain-uplift in Ferrite, Austenite, and Copper.

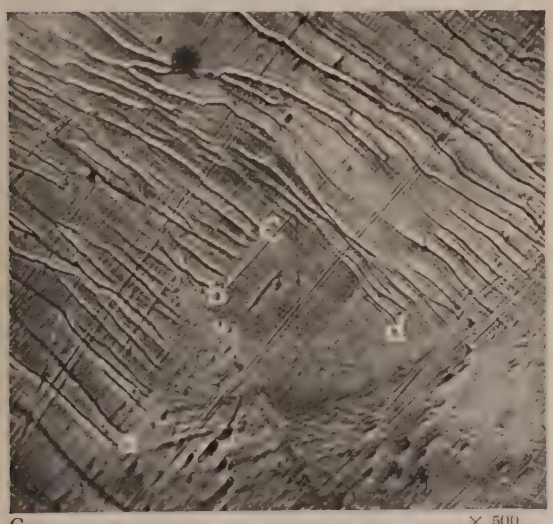


A. × 500

Slip-bands in twinned areas in copper. §499.

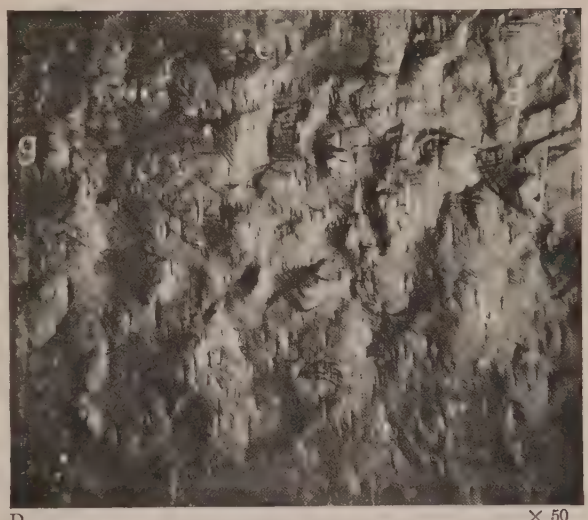
B. × 500

Slip-bands crossing grain boundaries in pre-etched ferrite. §425.



C. × 500

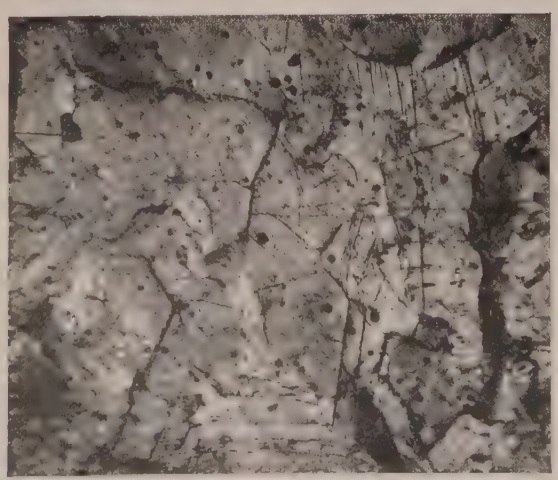
Slip-bands in unetched ferrite. §408.



D. × 50

Grain-uplift and slip-bands in austenite. §408.

E. × 50



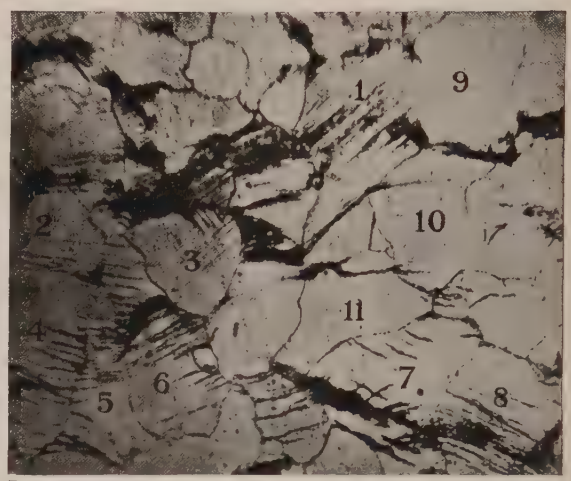
F. × 50

Grain-uplift and slip-bands in austenite. §§409, 412.

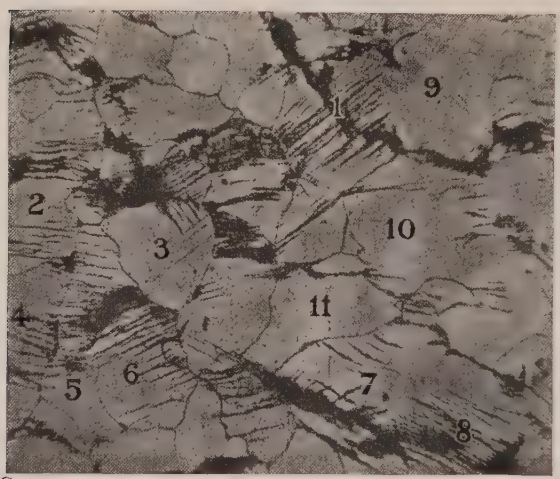
PLATE 25

Slip-bands from Reversal of Stress of 14.3 Tons per Sq. In.—*Ewing and Humfrey* (A to D). §§422, 497.

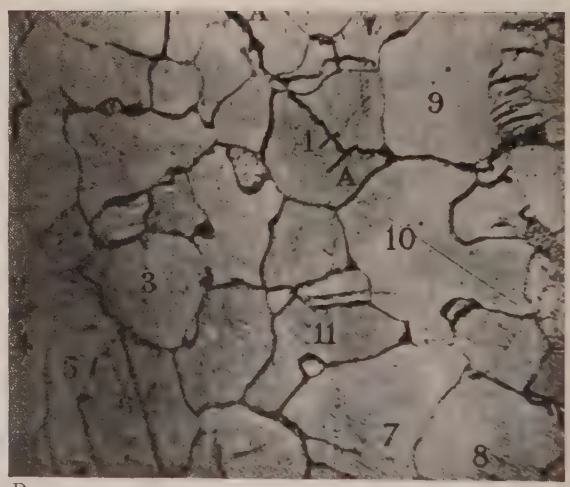
A. After 5000 reversals. × 150



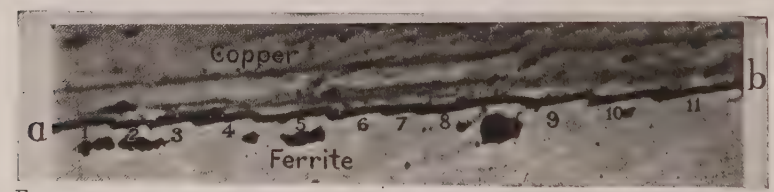
B. After 40,000 reversals. × 150



C. After 60,000 reversals. × 150

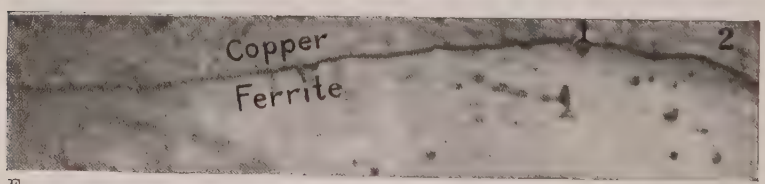


D. After 70.000 reversals, repolishing and etching. × 150

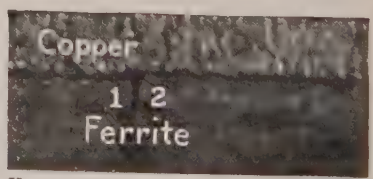


E. × 2000 —*Rosenhain*

E.



F. × 2200



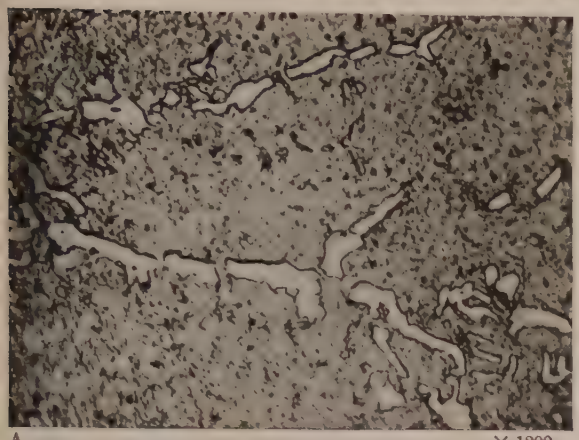
H. × 1200

G. × 1200

I. × 1200

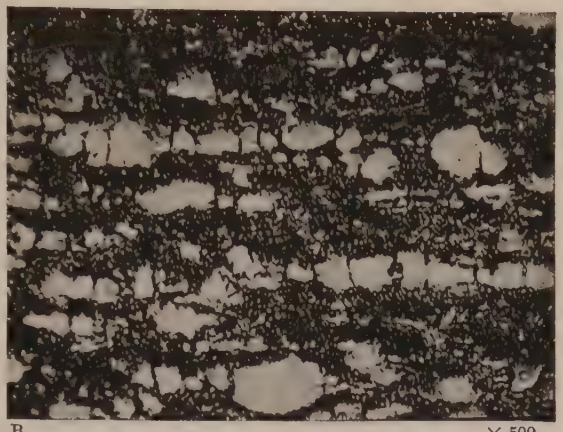
(E to I) Copper-plated cross-sections through slip-bands in ferrite. §§407, 418, 466.

Deformation in Carbon Steel and Wrought-Iron, (§§108, 680) and Silhouette.



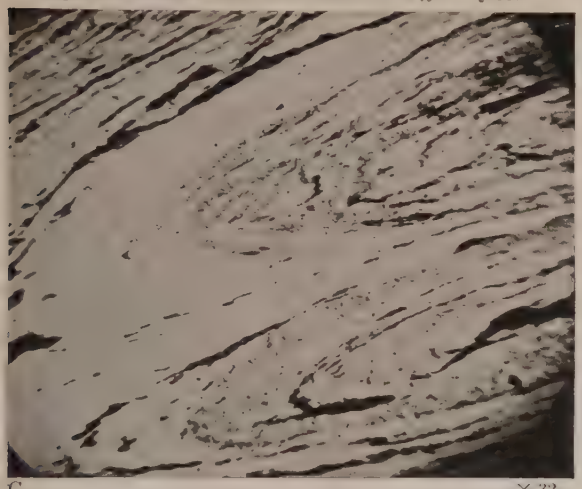
A. × 1200

Rupture of divorced cementite in steel of 0.21 % C. §539.



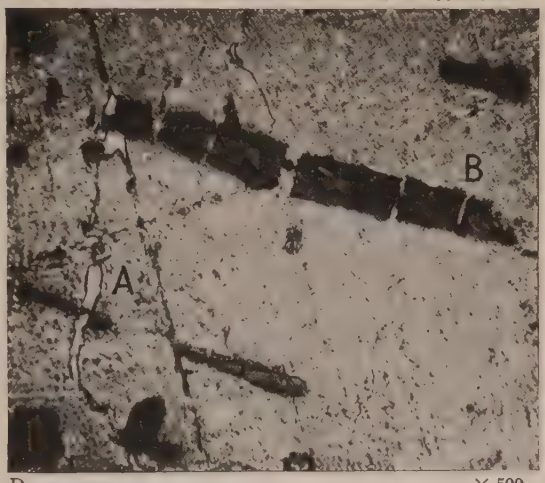
B. × 500

Rupture of cementite masses in steel of 2.04 % C. §§532, 537.



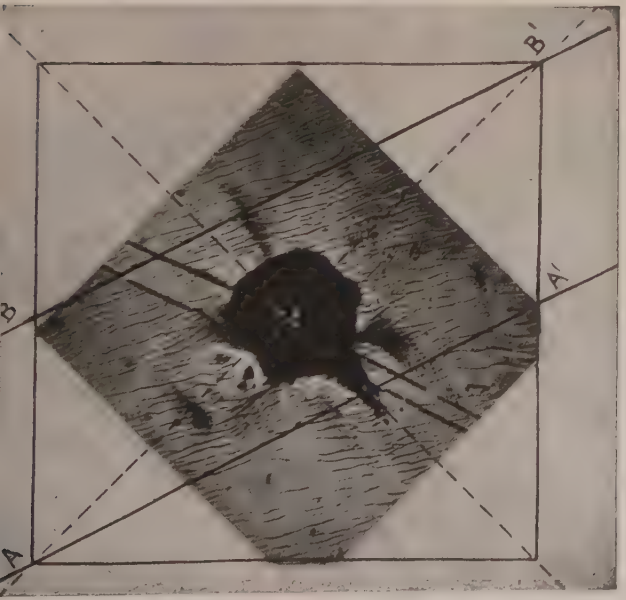
C. × 32

Flow and rupture of slag in bending wrought-iron while hot. §766.



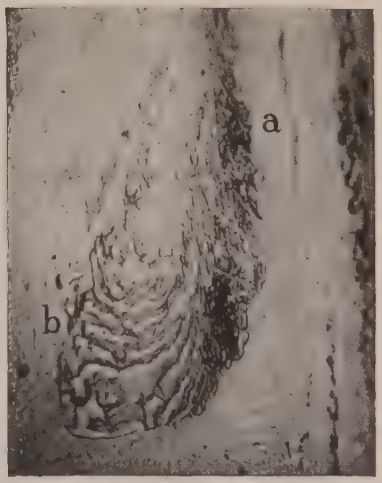
D. × 500

Rupture of slag and inflow of ferrite in punching wrought-iron. §721.



E. × 130

Irrelation between silhouette in ferrite and later slip-bands.—After *Osmond and Cartaud*. §431.

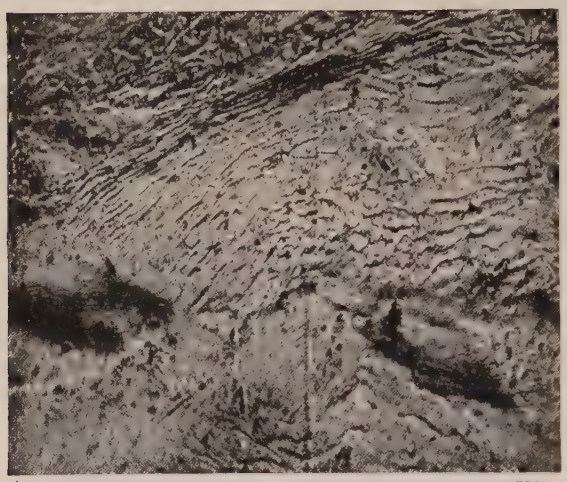


F. × 1200

Movement of pearlitic cementite sheets in punching.—*Howe and Levy*. §541.

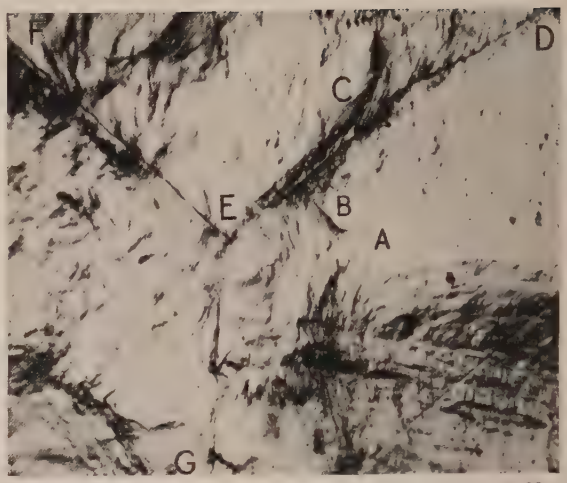
PLATE 27

Deformation of Carbon Steel.



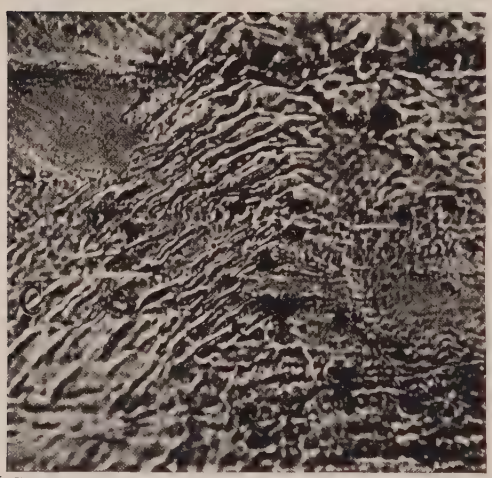
A. × 500

Extrusion of pearlitic ferrite by pressure. C = 0.92. §542.

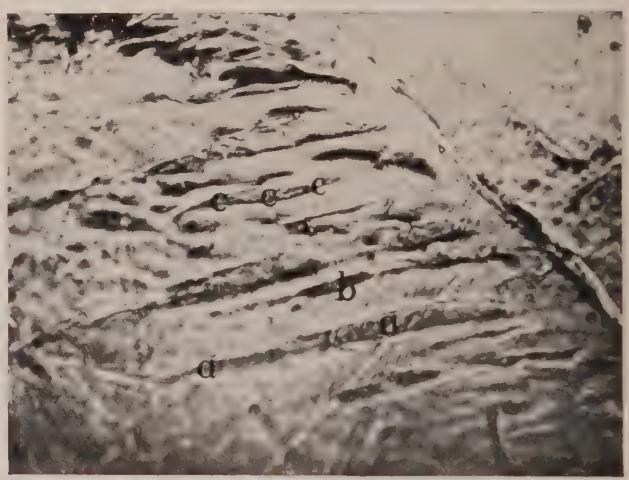


B. × 90

Extrusion of ferrite network GE and EF by pressure. C = 0.40. §§531, 542.



C. × 180



D. × 180

Deformation lines in sorbite simulating slip-bands. §543.

E. × 50

Extrusion of ferrite network by pressure. Oblique light. C = 0.59. §§531, 725.

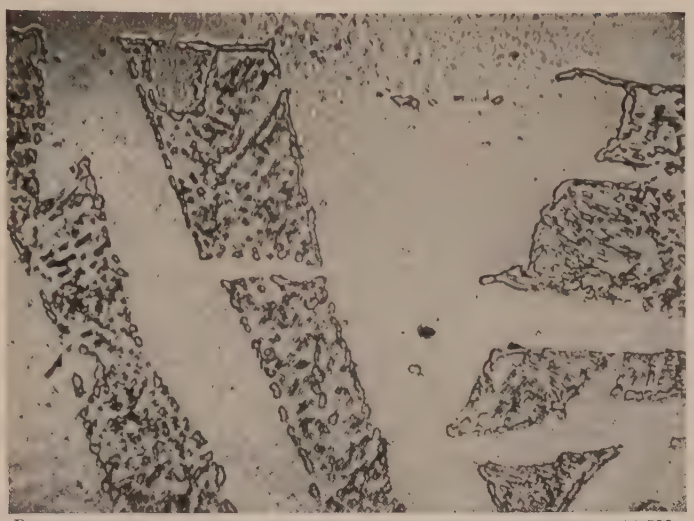
F. × 500

Extrusion of ferrite network *ab* and of pearlitic ferrite in steel of 0.59 C by pressure. Same as Fig. E. §§531, 543, 725.

Deformation in Carbon Steel, and Widmanstättian Figuring. (§§235, 531, 543.)

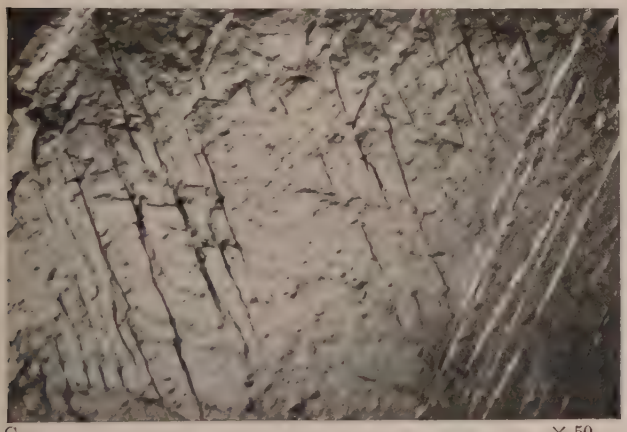
A. × 50

Widmanstättian structure of ferrite. C = 0.40.



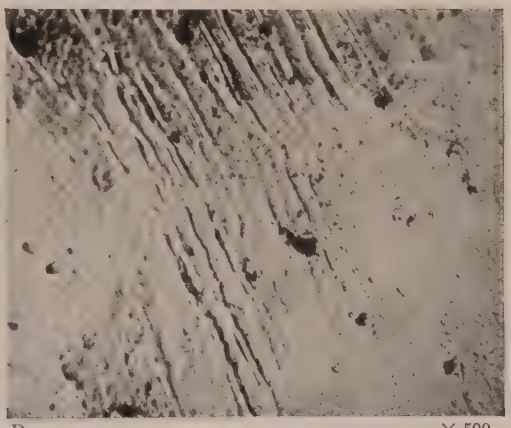
B. × 500

Undercooling cementite about the ferrite beams of Fig. A.



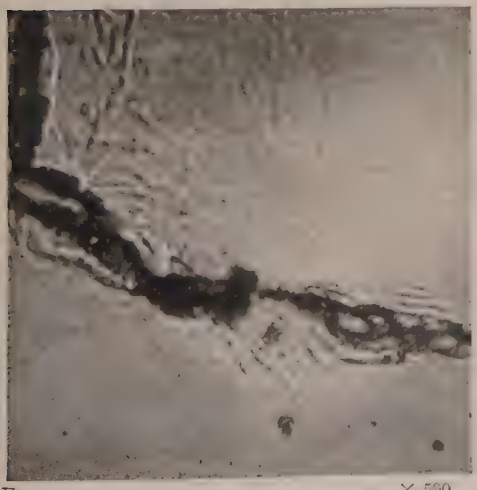
C. × 50

Extrusion of Widmanstättian ferrite beams by pressure. Unetched. C = 0.40.



D. × 500

Slip-bands in Widmanstättian ferrite bands of Fig. A. Unetched. C = 0.40.



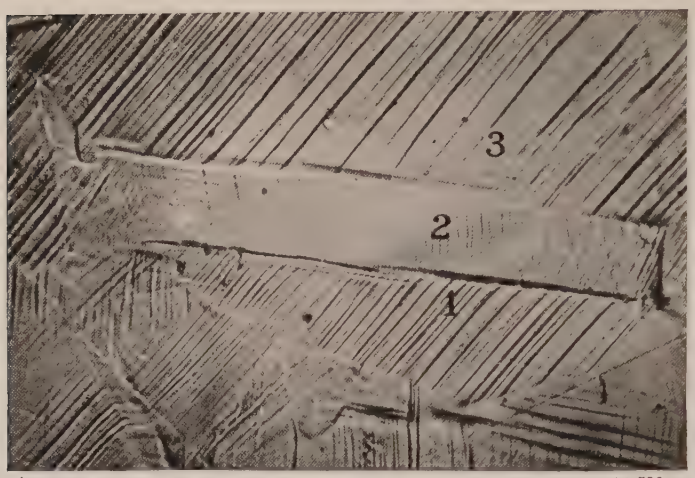
E. × 500

F. × 509

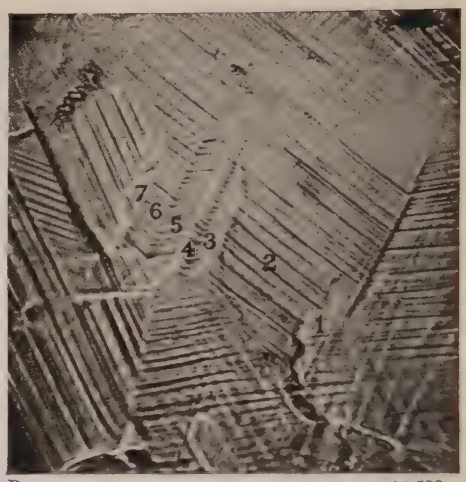
Extrusion of martensitic markings by pressure in hardened steel of 0.40 C. §545.

PLATE 29

Annealing Twins in Hadfield's Manganese Steel and Invar. Heated After Deformation, Polished, and Squeezed in Vise or Hammered Gently. §§558, 560.



A. × 500

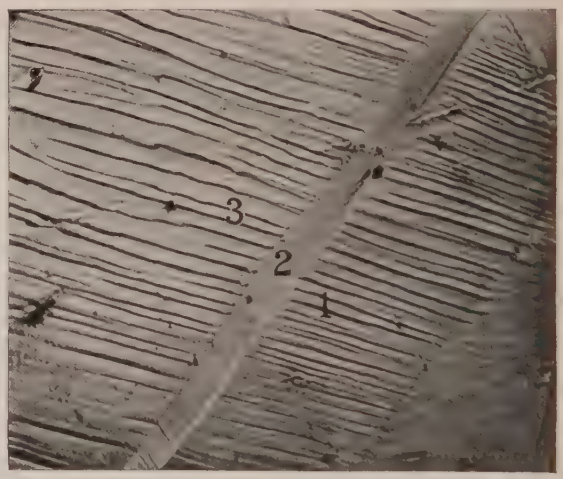


B. × 500

Near the fracture of a tensile test piece of austenitic quenched manganese steel. §412.

C. × 500

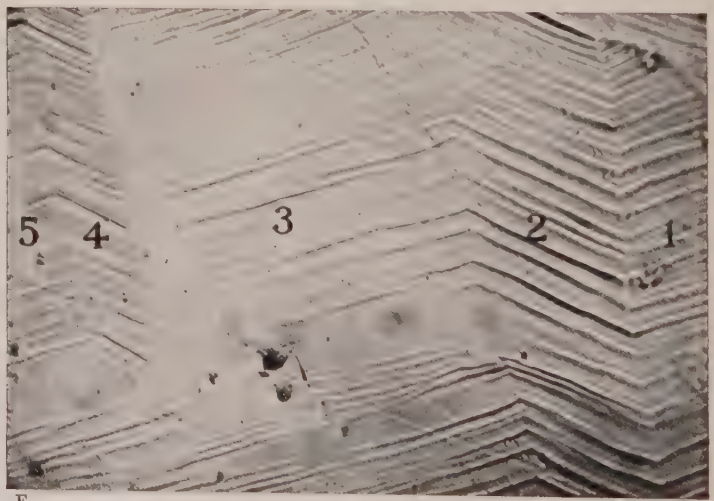
Austenitic manganese steel squeezed in vise. §566.



D. × 500

Invar, same treatment as Figs. E and F.

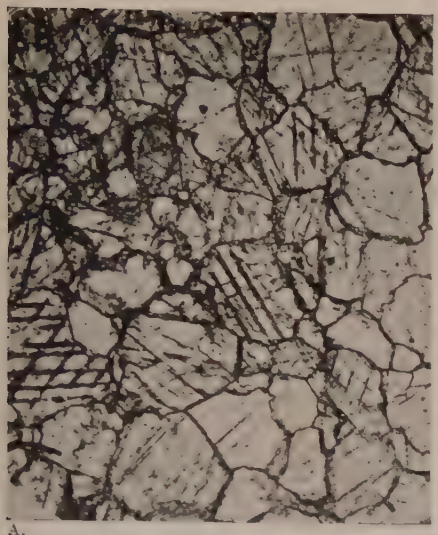
E. × 500



F. × 500

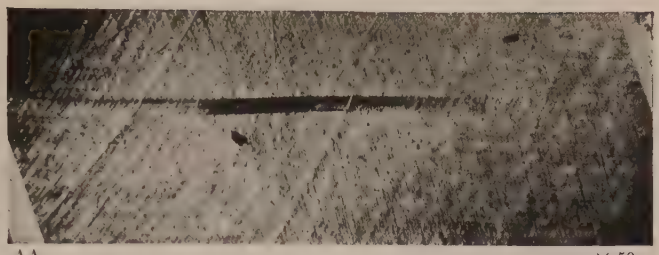
Invar, squeezed severely, annealed at 600°, polished, squeezed gently.

Twins In Copper, Austenite, and Minerals. Neumann Bands (?) in Ferrite.



A.

Mechanical twins in calcite.—*Adams and Nicholson.* §550.

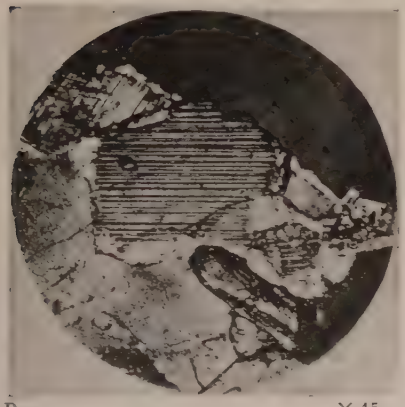


AA. × 50

Twin in native copper crystal very slightly deformed.

AB × 50

The same twin after more deformation. §§501, 558.



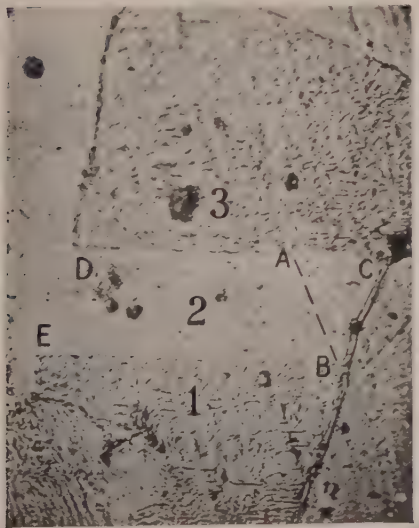
B. × 45

Twinning striation of plagioclase.—*Rosenbusch.* §551.

C. × 50

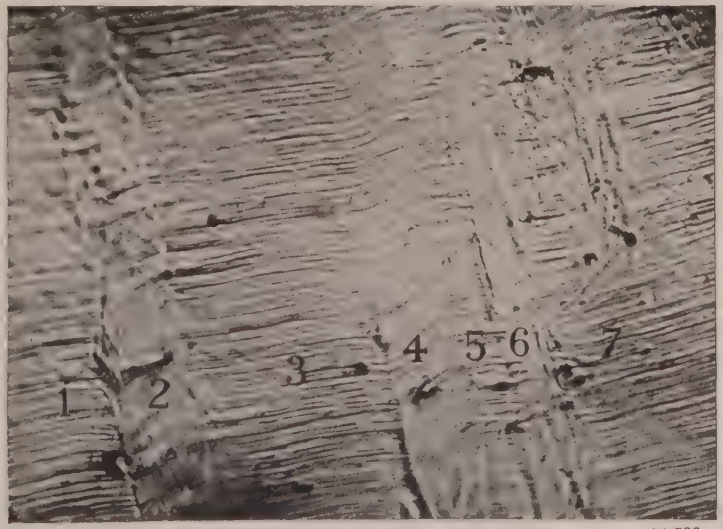
D. × 50

Reversal of relative brightness of twinned areas on reversing direction of light. §§546, 561.



E. × 500

Annealing twins in 33 per cent. Mn steel shown by etching. §§639, 613.



F. × 500

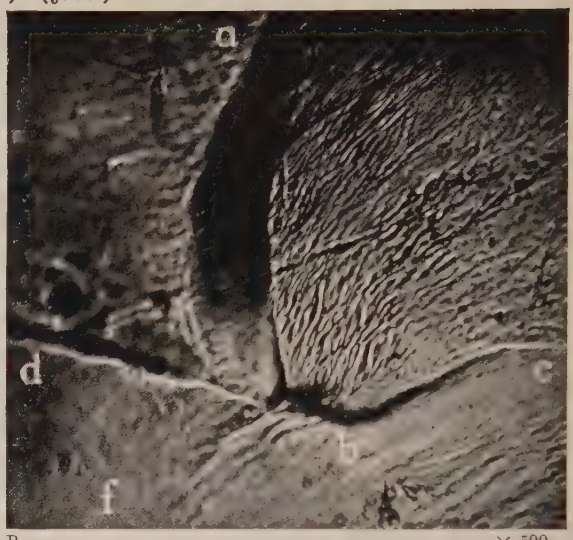
Neumann bands, 2, 4 and 6, formed mechanically in ferrite, which was next polished and squeezed gently. §§558, 600.

PLATE 31

Slip-bands and Neumann Bands Developed by Pressure Alone. (C was Etched Before the Development.) (§570.)

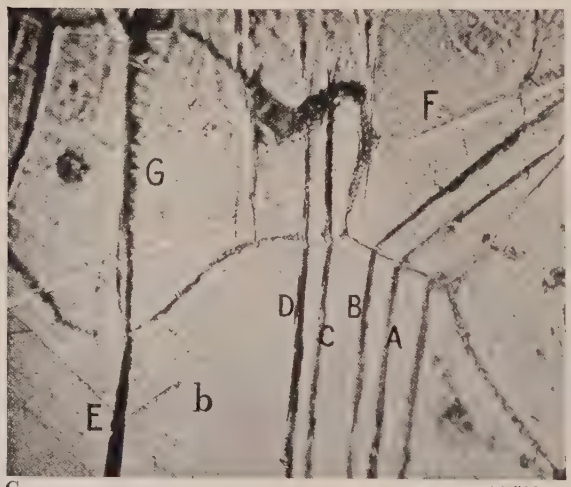
A. × 500

Curves in Neumann bands. §410.



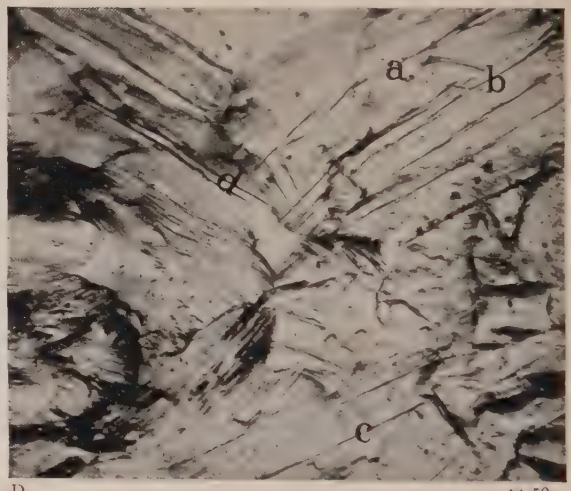
B. × 500

Slip-band projections in ferrite. §410.



C. × 500

Neumann bands deflect at grain boundaries. Pre-etched. §604.

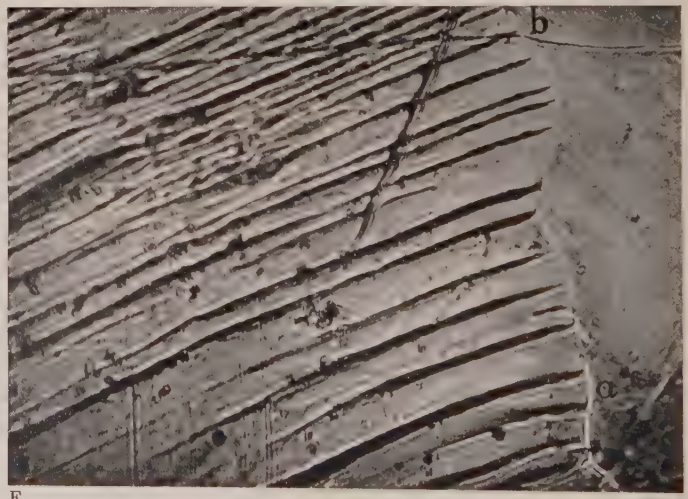


D. × 50

Neumann and slip-bands. §588.

E. × 50

Neumann bands under oblique light. §571.



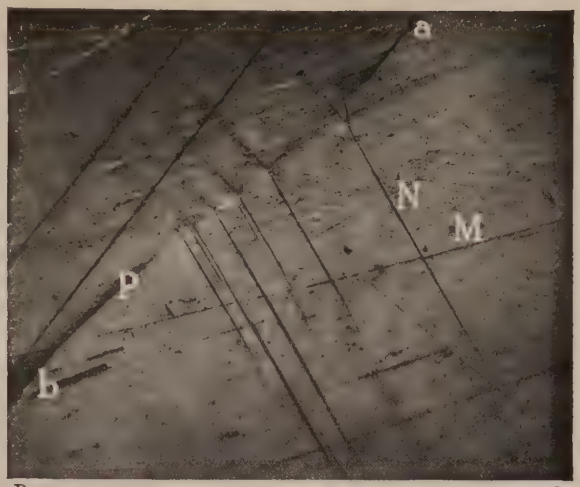
F. × 500

Neumann bands (?). §604.

Neumann Bands on Unetched Surfaces of Ferrite (Silicon Steel). (§570.)

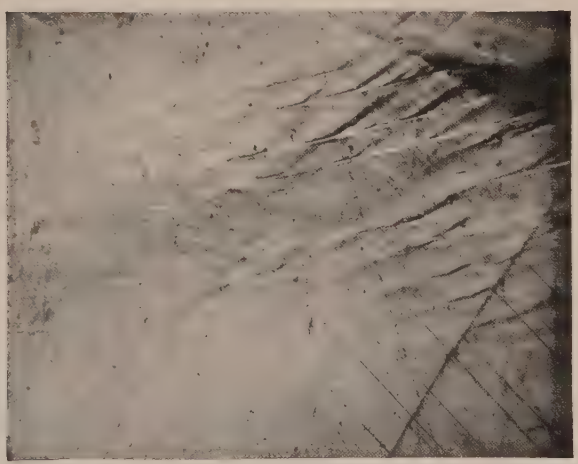
A. × 50

Grain uplift (?) §§409, 410.



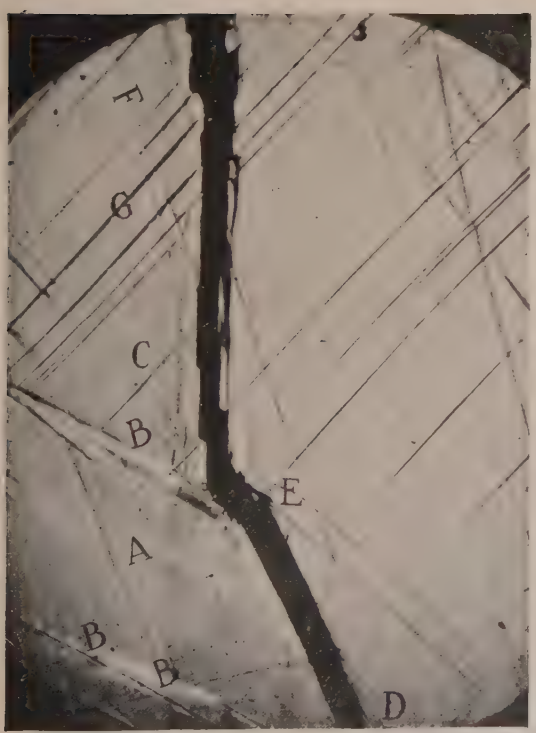
B. × 50

Curvature caused by faulting. §409.



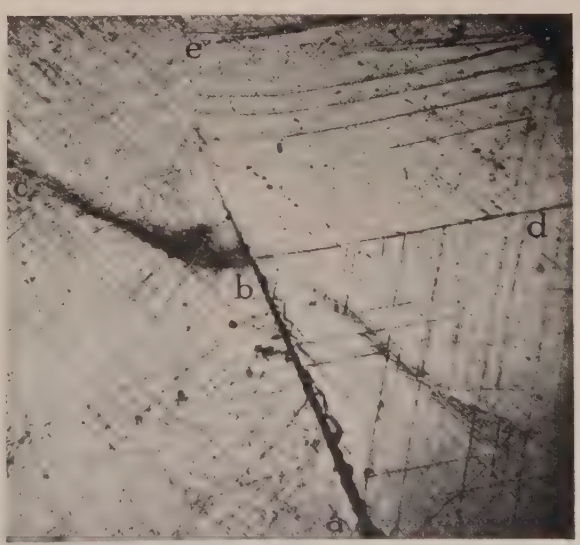
C. × 50

Intersecting Neumanns.



D. × 50

Rupture follows Neumann directions. §603.

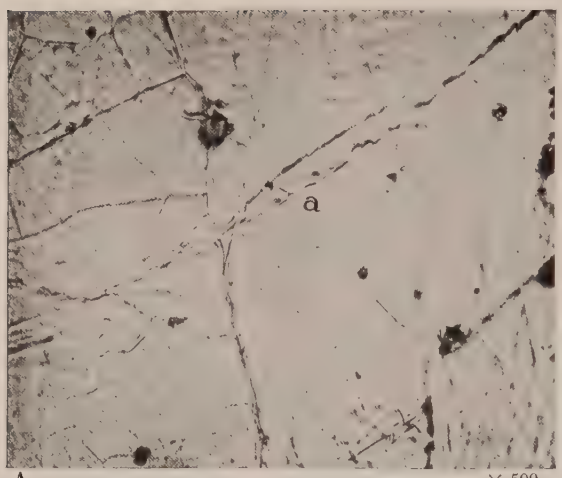


E. × 50

Curved Neumanns. §409.

PLATE 33

Neumann Bands Developed by Etching. (§§570 *et seq.*)

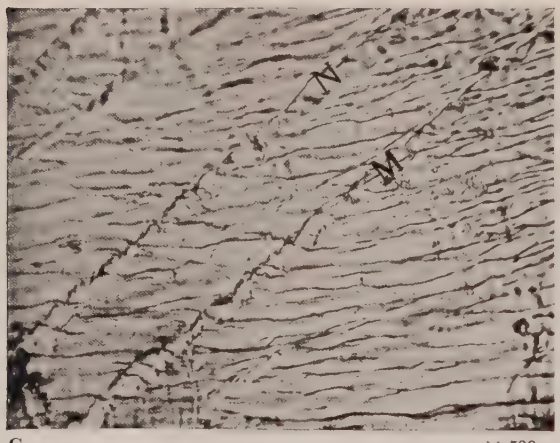


A. × 500

Zigzagging of slip band a on crossing Neumann. §600.

B. × 500

Ladder pattern. §578.

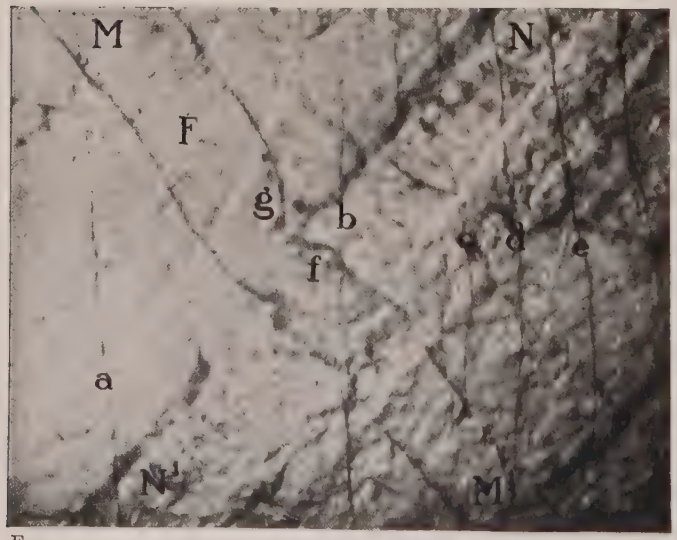


C. × 500

Faulting of Neumann about slip-bands. §576.

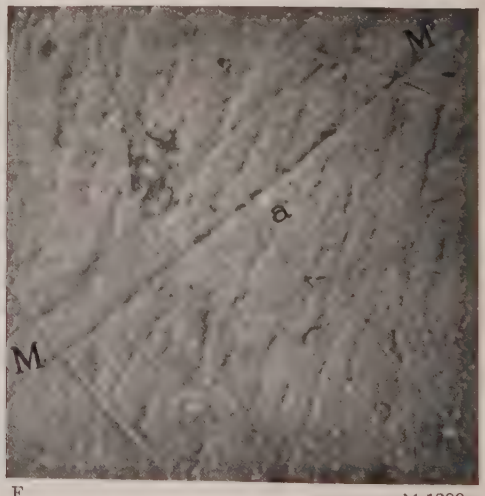
D. × 500

Neumann MM′ faults NN′. §588.



E. × 1200

Crossing of Neumann (?) and slip-bands. §§573, 588.



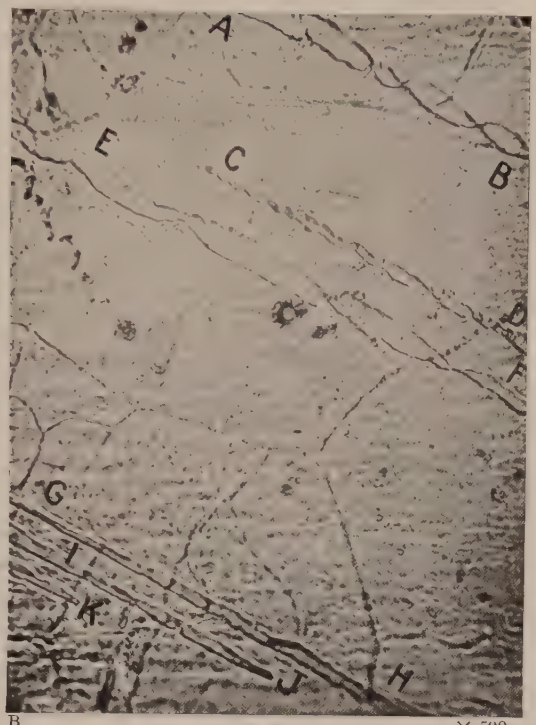
F. × 1200

Discontinuous slip-bands (?) §424.

Neumann Bands Developed by Etching.

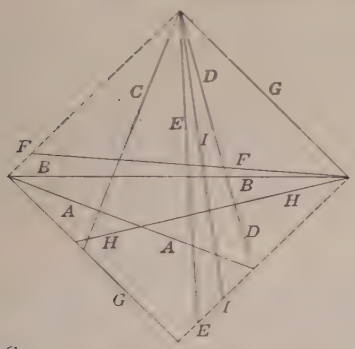
A. × 50

8 directions for Neumanns on cube face. §590.



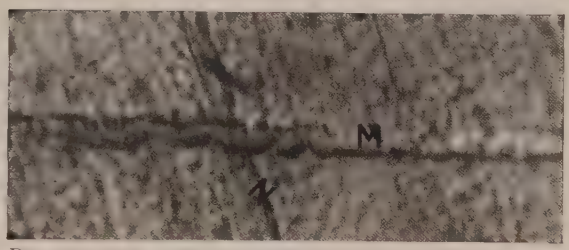
B. × 500

Nodal Neumanns. §§575, 604.



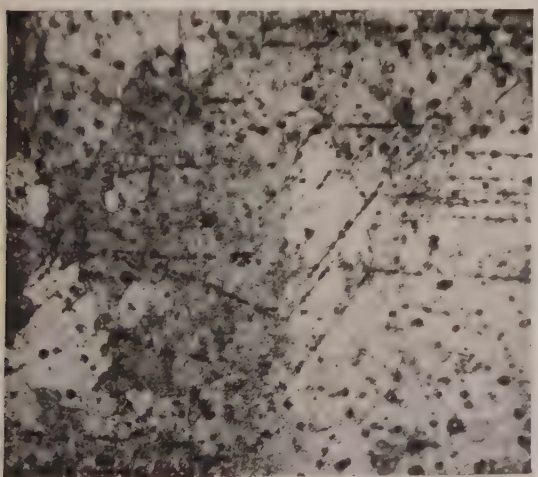
C.

Key showing crystallographic directions of Neumanns in Fig. A. Cube face turned 45° from conventional position.



D. × 1200

Crystallographic serrations. §605.



E. × 50

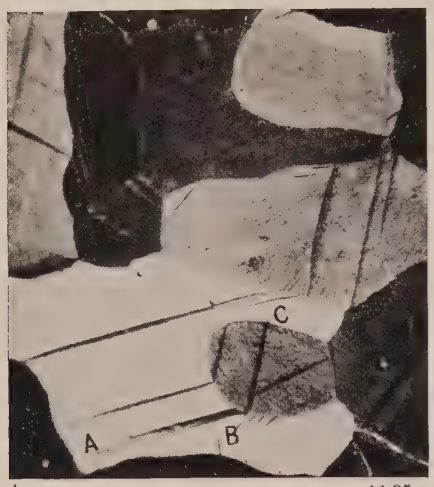
Spotty widening of Neumanns. §572.

F. × 500

6 directions of Neumanns on cube face. Detail of Fig. A. §602.

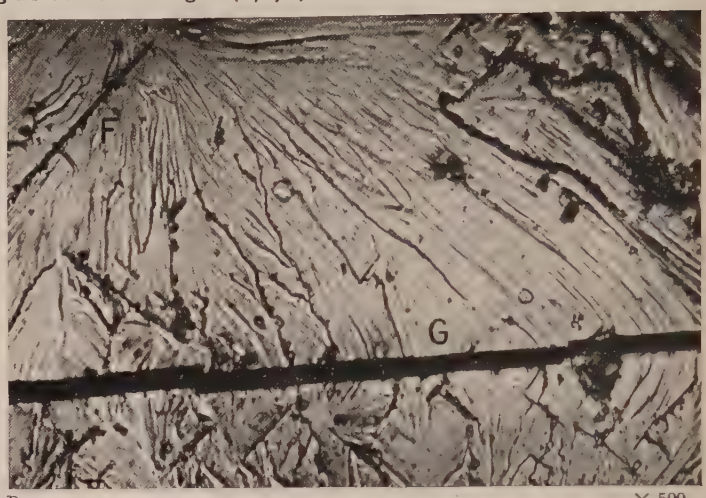
PLATE 35

Neumann Bands on Pre-etched Surfaces (A), on Cubic Cleavages (B,C,D), and Deepened by Etching in Cubic Cleavages (E,F,G).



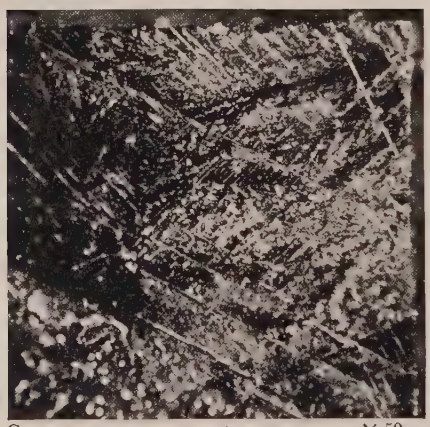
A. × 35

Neumanns in etched cast zinc.—*Campbell.* §604.



B. × 500

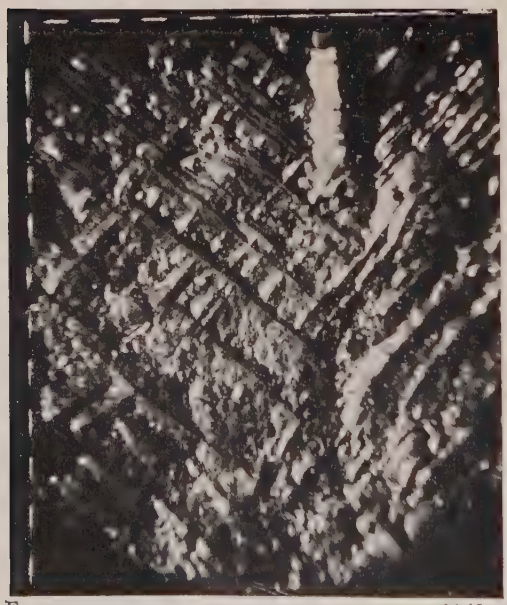
Cubic and 211 Neumanns on cube face. §§587, 590.



C. × 50

D. × 90

Neumanns on 2 cube faces of Fig. B. Oblique light. §§587, 590.



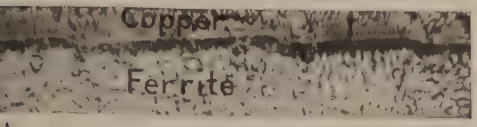
E. × 10

F. × 10

G. × 10

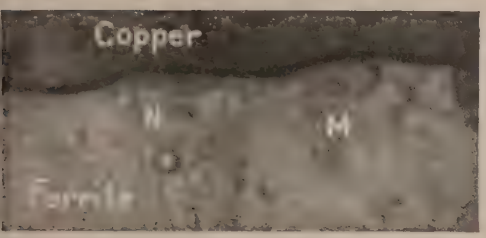
Deeply etched Neumanns on 3 adjacent cube faces of one prism. §§371, 590.

Profiles of Neumann Bands (A to E), Neumann Bands Developed by Etching (F,H,I), and X-Bands (G).

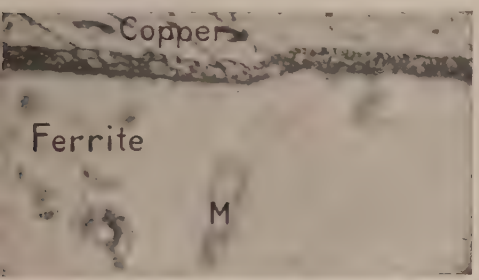


A. × 180

B. × 1200



C. × 1200

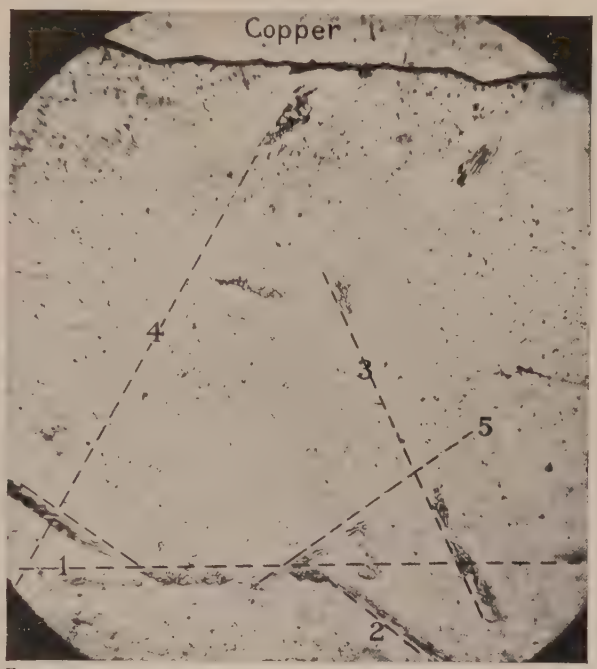


D. × 1200

Copper-plated profiles of Neumanns.
§§466, 579, 602, 612.

E.

Profile of etching furrows of Fig. E., Plate 35. §590.

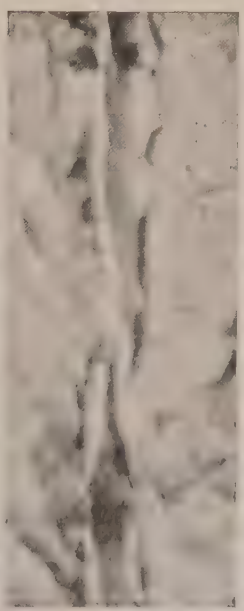


F. × 50

Neumanns follow 211 directions on 011 faces. §590.

G. × 500

X-bands in punched ferrite. §626.



H. × 1200

I. × 500

Details of Neumanns of Fig. F. §575

PLATE 37

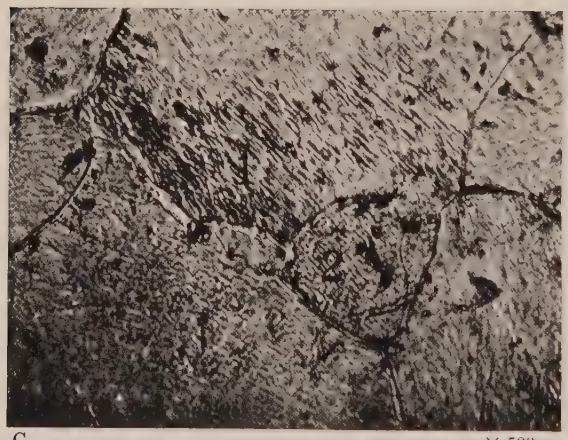
X-Bands in Freshly and Deeply Etched Ferrite After 30,000 lb. Compression per sq. in. (A to E) or Punchi (F and G), and Later Treatment Given Under the Several Figures. (§§626, 635.)

A. × 500

After 2 hours rest in the cold.

B. × 500

After 49 hours rest in the cold.



C. × 500

After 38 days rest in the cold.

D. × 500

After 15 minutes at 100°.

E. × 500

After 15 minutes at 300°.

F. × 500

After 15 minutes at 200°.

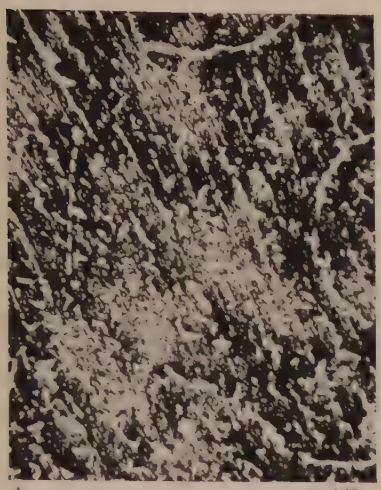
G. × 500

After 15 minutes at 600°.

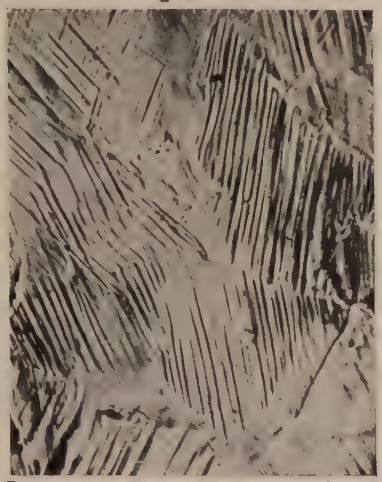
Deformation Lines in Manganese Steel and Copper. (§§412, 640, 646, 653, 655.)

Progress of the Development of Slip-bands.

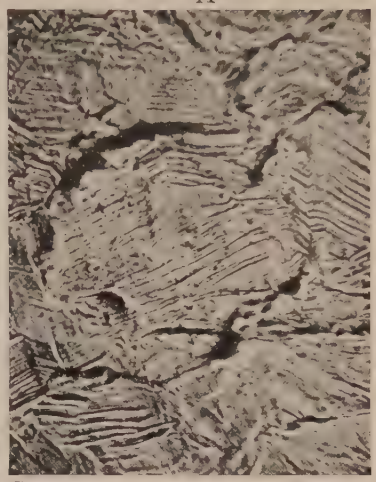
In Manganese Steel. In Copper.



A. × 180
Cementite precipitated by heating to 750°.



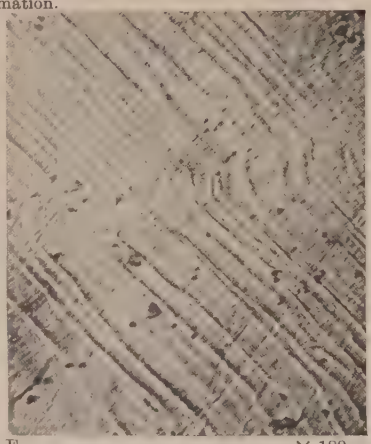
B. × 500



C. × 500
Great deformation.

D. × 500
Cementite precipitated by heating to 575°.

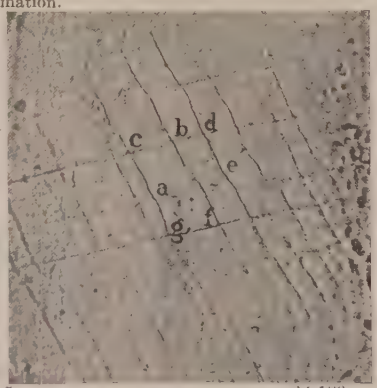
E. × 180



F. × 180
Less deformation.

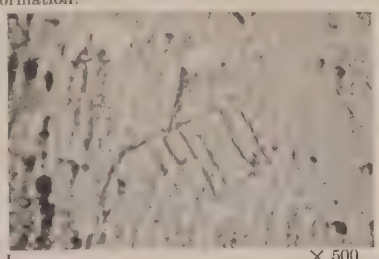
G.
Eutectiform pattern related to dendrites.

H. × 180



I. × 180
Still less deformation.

K. × 500



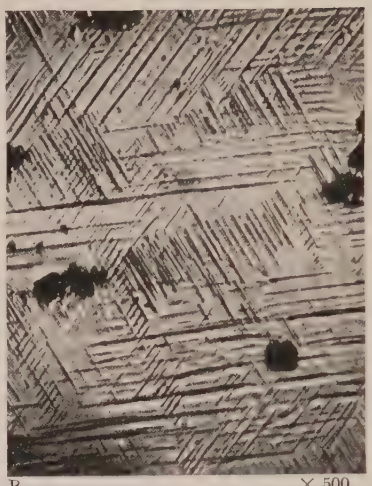
L. × 500
Least deformation.

PLATE 39

Deformation Lines in Quenched Manganese Steel. (§§412, 641, 645, 656, 657.)

A. × 500

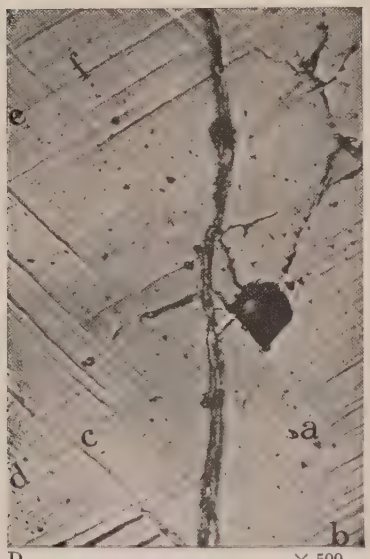
Etching lines without heating.



B. × 500

C. × 500

Martensitic structure after complete (B) and partial (C) removal of etching pellicle.

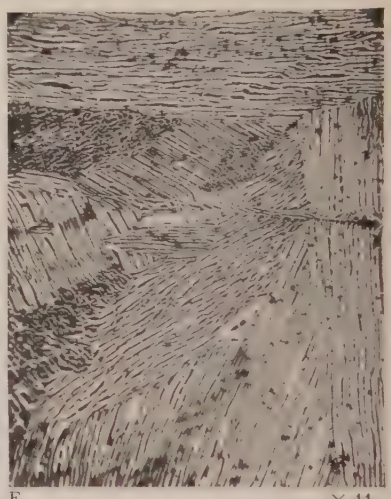


D. × 500

Conjugate slip-bands suggesting twinning.

E. × 500

Slip-bands stop or change prominence at boundaries.

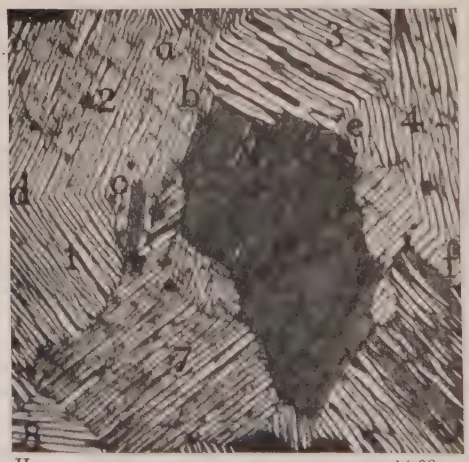


F. × 44

Columnar eutectiform pattern.

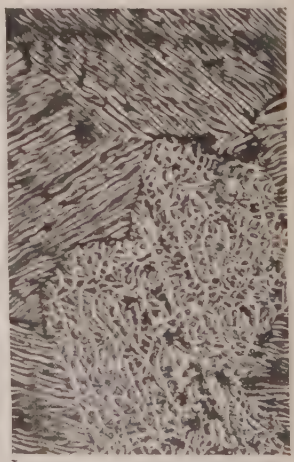
G. × 50

Slip-bands cross dendrites and grain boundaries.



H. × 90

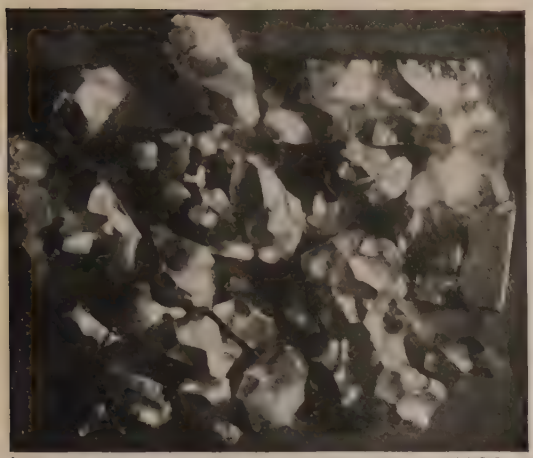
Eutectiform pattern.



I. × 65

Crackly eutectiform pattern.

Fractures in Iron, and Path of Rupture in Bismuth.



A. × 0.9

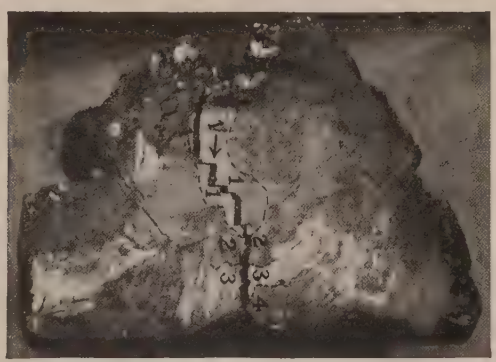
Curved intergranular chondritic fracture of silicon steel. §692.

B. × 0.22

Cementite plates in fracture of spiegeleisen. §731.

C. × 0.3

Barked wrought iron. §762.

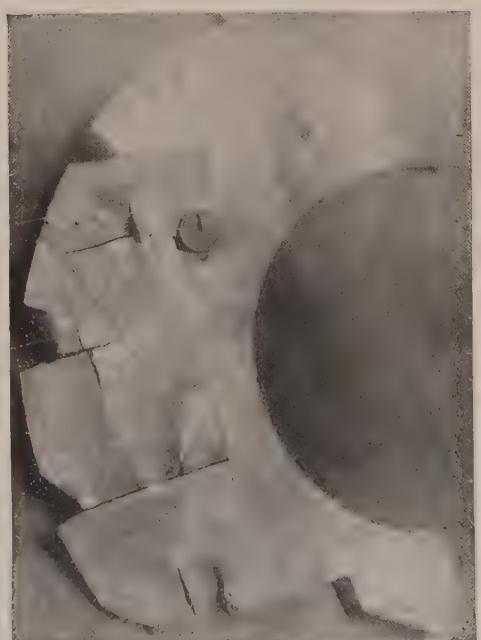


D. × 2.5

Rupture seeks grain centres in bismuth. Etched. §694.

E. × 50

Rectangular etching figures and rupture in silicon steel.—STEAD. §738.

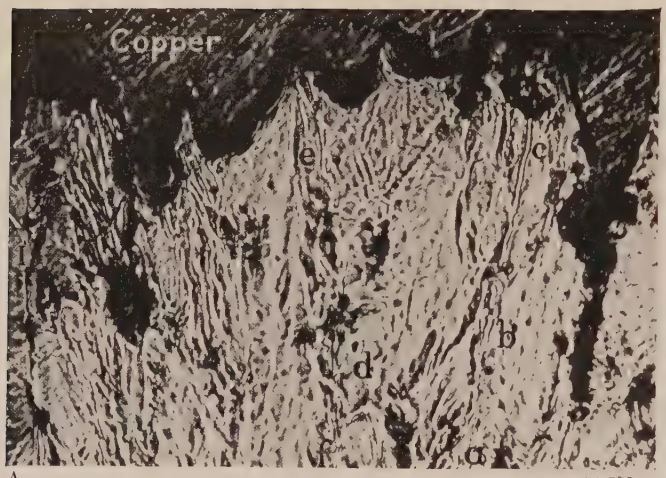


F. × 0.5

Rectangular fracture in low-carbon sheets.—*Stead.* §740.

PLATE 41

Path of Rupture in Steel, Wrought-Iron, and Bismuth.



A. × 500

Copper-plated fracture of ferrite. §724.

B. × 30

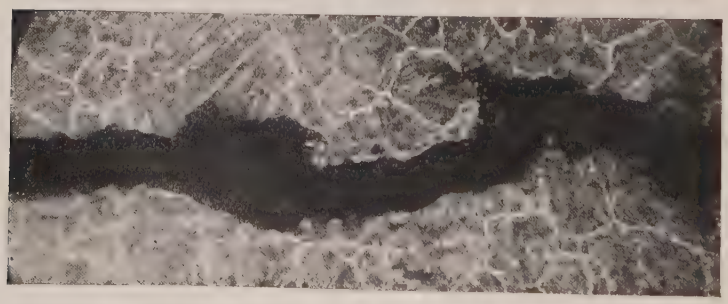
Rupture follows slag in wrought-iron. §764.

C. × 50

Rupture escapes ferrite network. C = 0.59. §§717, 725.

D. × 50

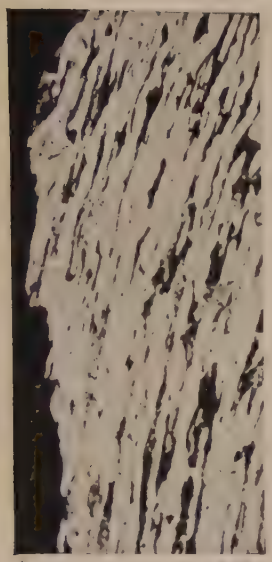
Trans-crystalline rupture, Neumann bands, and broad twins, O, in bismuth. §§570, 609, 685.



E. × 50

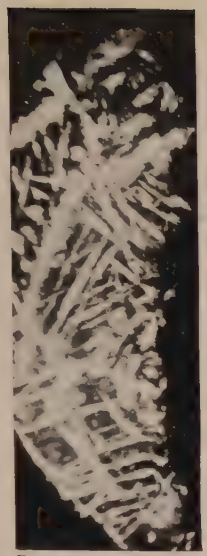
Rupture follows cementite network in hyper-eutectoid steel. §731.

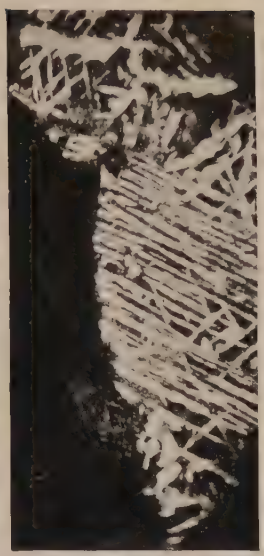
Path of Rupture.



A. × 66

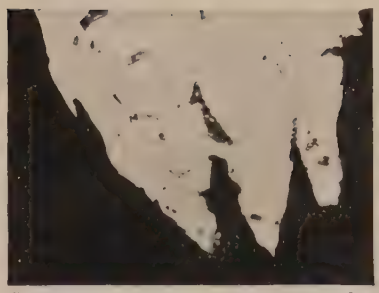
Rupture in 0.14 C steel. §717.





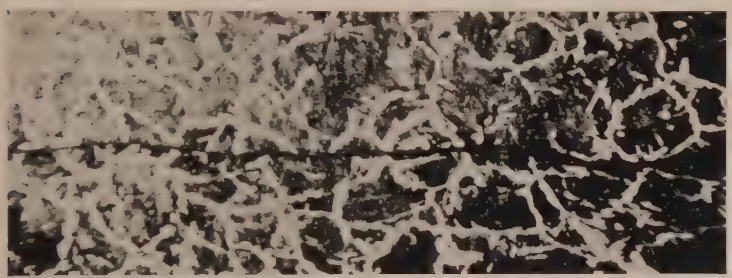
B. × 80

Rupture escapes ferrite in Widmanstättian structure. §719.



C. × 250

Intergranular rupture at 932°. C = 0.106.
Rosenhain and Humfrey. §689.

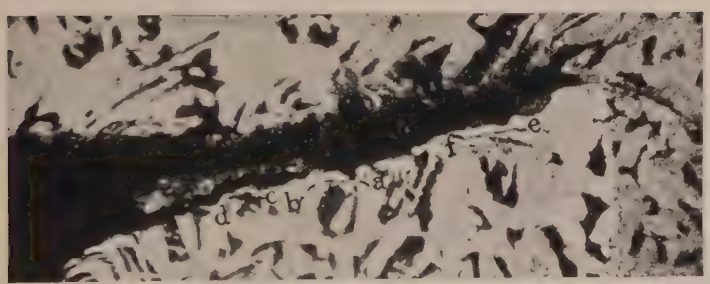


D. × 50

Indiscriminate path of rupture in rail steel. §720.

E. × 500

Rupture follows intergranular cementite. §731.

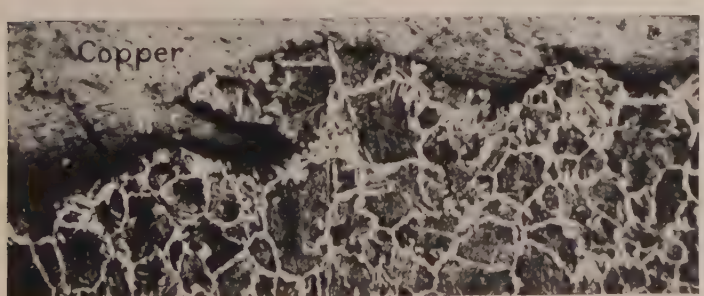


F. × 50

Rupture avoids ferrite. C = 0.204. §717.

G. × 1.4

Fibrous fracture in nickel. §762.



H. × 50

Little discrimination in rupture. C = 0.46. §720.

I. × 1.4

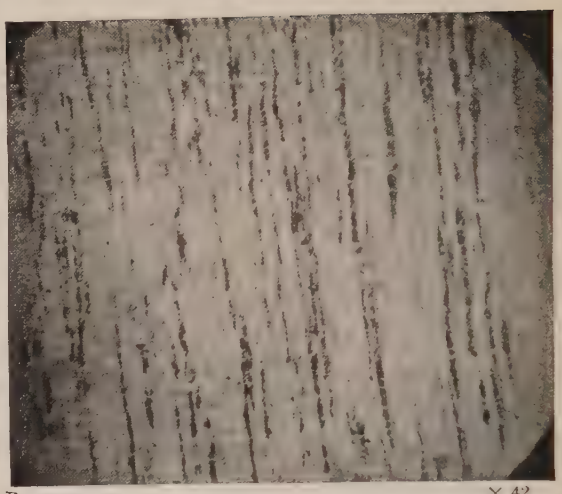
Fibrous fracture in nickel steel. §§749, 762.

PLATE 43

Ferrite Ghosts. (§§367, 762, 780, 786.)

A. × 42

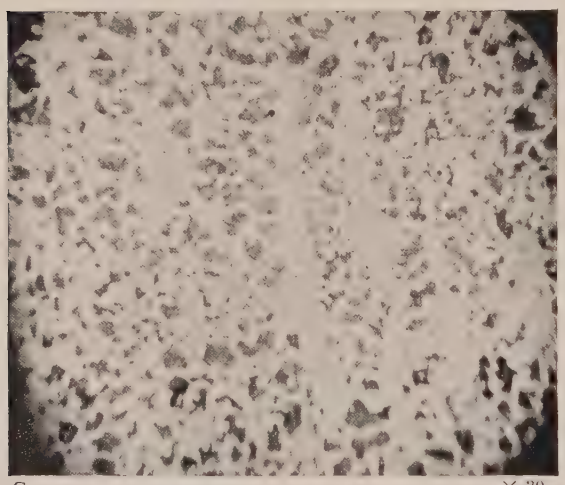
Air cooled.



B. × 42

Furnace cooled.

On cooling from 1000°.

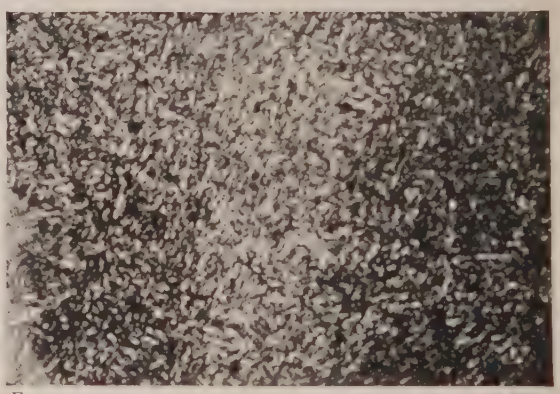


C. × 30

Phosphorus steel, P = 0.417, C = 0.34.

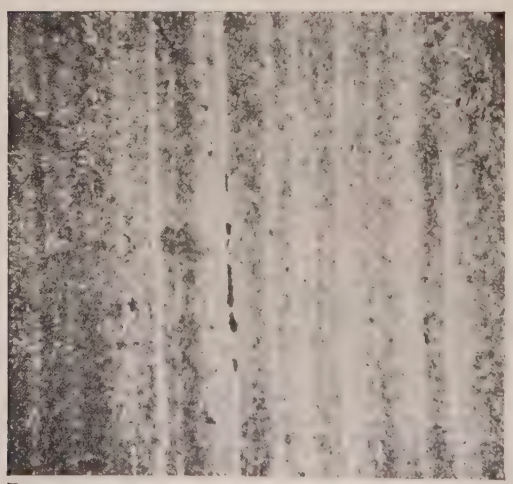
D. × 0.95

Macroscopic ghosts.
C = 0.46, P = 0.041



E. × 500

Seen best from a distance. C = 0.40, P = 0.014. Reheated to 600° after quenching from 1100°.



F. × 50

On reheating to 600° after quenching from 800°.
C = 0.46, P = 0.041.

G. × 50

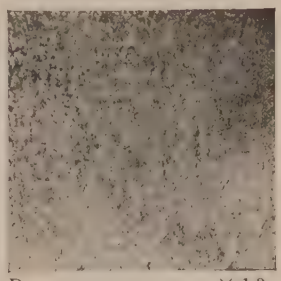
Ghosts in steel of 0.014 P.

A to F, Boiler Plates. A, B, C, $\frac{3}{8} \pm$ in., D, $1\frac{1}{4}$ in. thick, Rolled From 13 in. thick Ingots. A, B, and C Sheared; D Universal. §§(762, 809.) G to I, Miscellaneous.

A. × 1.3

B. × 1.3

C. × 1.3



D. × 1.3

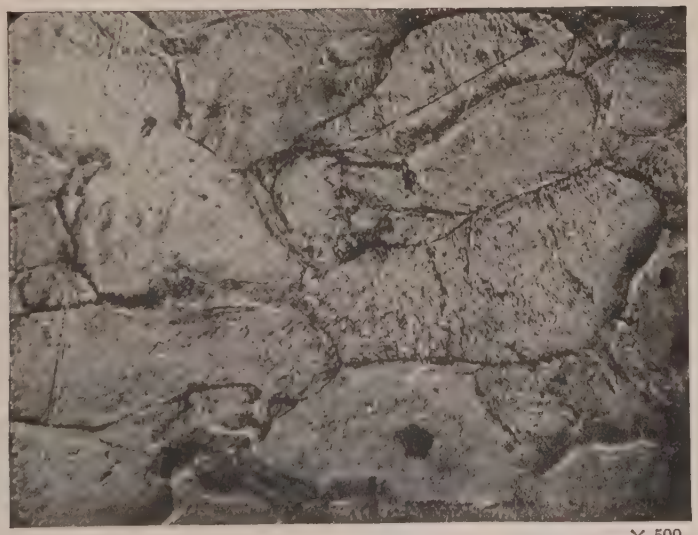
Direction of rolling.
Last period. ↓ → ↓ → ↓ → ↓ ↓
Preceding period.

E. × 28
Transverse of D.

F. × 50
Longitudinal of B.

G. × 500
Slag in ferrite. C = 0.43 §§367, 766.

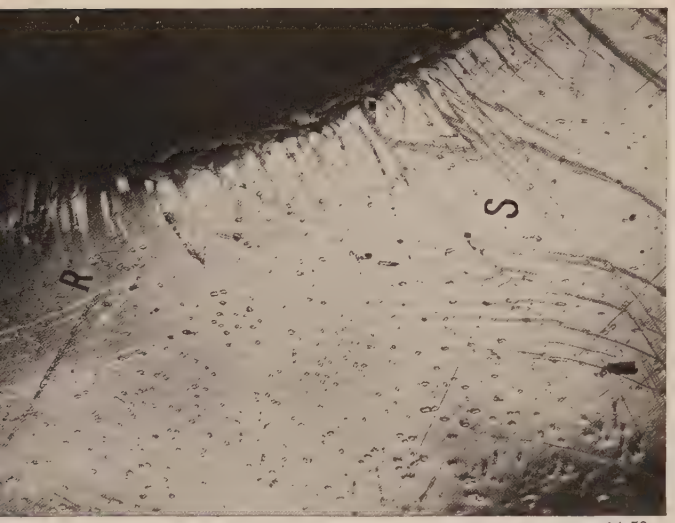
H. × 50
Combination of ghosts and network structure C, = 0.46. §796.



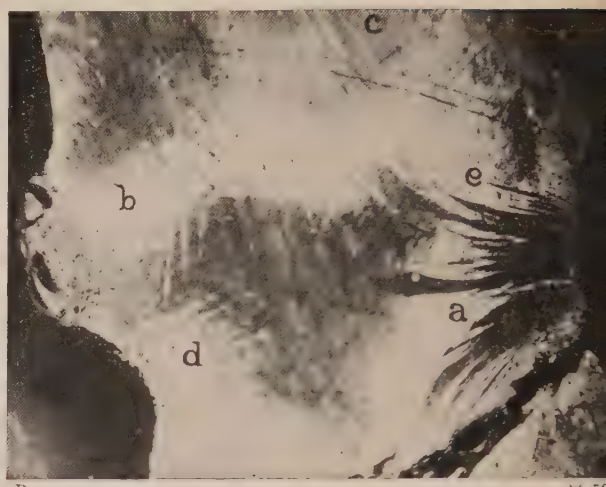
I. × 500
Ferrite grains deformed by 50% reduction by cold rolling. §§680, 765.

PLATE 45

Deformation Lines in Twisted Crystals (A, B, D, E, F) and Thorns and Edgings (C).



× 50

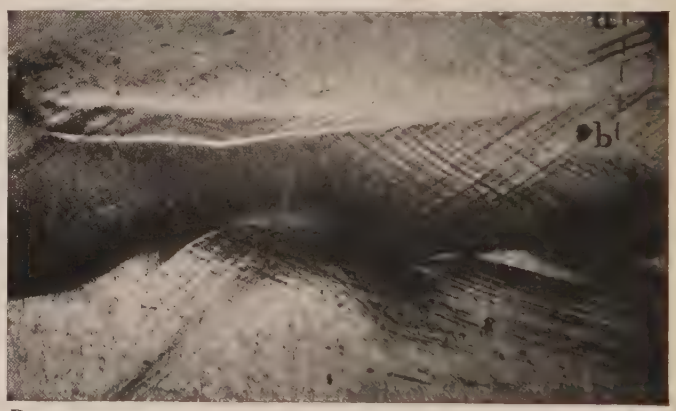


B. × 50

Fringes and Neumanns in twisted meteorite.—STEAD. §§582, 599.

× 195

Thorns (Q, R, S) and edgings (T) in ferrite.—*Osmond, Fremont, and Cartaud.* §583.



D. × 50

Geometrical slip-bands in twisted copper crystal. Longitudinal section. §§501, 582, 676.

× 50

Geometrical slip-bands in copper crystal. §§412, 417, 441, 501.

F. × 50

Straight twin in twisted copper crystal. Cross section. §§582, 676.